METALLURGIA

The British Journal of Metals

(INCORPORATING THE METALLURGICAL ENGINEER)

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PRINCIPAL CONTENTS IN THIS ISSUE:

	Page
Mutual Aid for Freedom	1
Planning of the Gas Industry	2
The Weldability of Cast Iron. By T. J. Palmer, A.M.Inst.W. .. .. *A welding process where the heat input per unit volume is relatively low is considered to be most suitable for joining cast iron. Particular attention is given to fusion welding, and the outstanding points relating to procedure are reviewed.*	3–6
Jigs and Tools for Anodising. By G. O. Taylor, M.S.E., A.I.P.E. .. .. *Further consideration is given to the design of jigs and tools for the handling of light metal articles during the process of anodising. Several tools are described which can be advantageously added to present equipment for this purpose.*	7–10
Difficulties Encountered in the Heat-treatment of Drop Forgings. By Bernard Thomas, F.Inst.P. .. *Many difficulties encountered by the heat-treater may be due to lack of control of heating in the mill or in the forge shop, but in this consideration of heat-treatment difficulties the author assumes that the drop-forging steel is in proper condition.*	11–14
The Late Dr. W. H. Hatfield, F.R.S. .. *An appreciation.*	15–16
The Use of Basic Lined Ladles in the Desulphurisation of Cast Iron by Sodium Carbonate. By N. L. Evans, B.Sc., A.I.C.	17–19
Correspondence .. .. *The Training of Female Analysts for Steelworks Laboratories.*	19–20
High-Silicon Acid-Resisting Cast Iron. By J. E. Hurst, D.Met. .. .. *The acid resistance properties of high-silicon cast iron are discussed and illustrated by diagrams. Attention is directed to its mechanical properties, and the manufacture of high-silicon iron components is briefly considered.*	21–25
Leaded Manganese-Molybdenum Steel. By T. Swinden, D.Met. .. .. *Tests have been carried out on material from a non-leaded and a leaded ingot of manganese-molybdenum steel in order to determine the effect of lead (0·19%) on the mechanical properties, including the machinability; grain size, and hardenability are also compared. The main results are given.*	26–30
The Structure and Segregation of Two Ingots of Ingot Iron, one Containing Lead. By L. Northcott, D.Sc., Ph.D., and D. McLean, B.Sc. ..	30–32
Metallurgical Investigations of Some Light Alloy Pistons from German Aircraft. By C. Wilson, Ph.D., A.R.S.M. .. .. *A summary of investigations is given.*	33–36
Electrolytic Production of Cobalt .. *For many purposes cobalt is required in a purer form than is possible by thermal reduction, and much attention has been given to electrolytic reduction; some of this work is reviewed.*	37–40
Palladium: Its History and Properties. By L. Sanderson	41–42
Metallurgical Digest.. .. .. .. *A new feature, presenting a digest of some foreign literature on various subjects within this journal's field.*	43–50

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135

METALLURGIA

THE BRITISH JOURNAL OF METALS.

INCORPORATING "THE METALLURGICAL ENGINEER."

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Mutual Aid for Freedom

WHEN the lend-lease arrangements were arranged with the United States, Britain was fighting with her back to the wall and with the support of the Nations of the British Commonwealth. At that time we badly needed war equipment and supplies to enable us to replace the great loss suffered at Dunkirk and thus enable us to continue the fight. The United States had been assisting us with supplies on a cash-and-carry principle, but the change to lend-lease made by the United States was a gesture of great practical importance, and there can be very few people in Britain who do not appreciate the enormous value of the help that has accrued as a result of that agreement. Whatever may be said of the present position existing between the two countries, therefore, will not detract in any way from the invaluable support this country received from the United States in the early days of the war.

There was, of course, much criticism of the lend-lease agreement in the United States, and criticism has shown a tendency to grow, largely due to the paucity of information on the subject available to the American people. Indeed, a large proportion of them were under the impression that the United States was continuing to give without return of any kind. That reciprocity has been in operation almost since the beginning of the agreement is well known to many in this country, and it seems surprising that no proper appreciation of this so-called reverse lend-lease has shown itself in the United States. However, there can be little doubt that people both here and abroad will be surprised with the information published in the Treasury White Paper. From this it is apparent that the material contribution made by the British Commonwealth under the reverse lend-lease system has become comparable in magnitude with the lend-lease supplies of the United States.

Last year President Roosevelt suggested that the costs of the war should be shared according to the respective capacity of each nation to contribute to the common pool, that, in fact, mutual aid to the maximum should govern the war partnership of the respective nations. It is noteworthy, therefore, that, according to the White Paper, what we give to the United States as mutual aid, in relation to our notional income, does not fall far short of what the United States gives to us as lend-lease, in relation to their national income. This is supported by President Roosevelt in his recent quarterly message to Congress on the progress of the lend-lease system, in which he states that British aid to the United States is now flowing at a rate of over £300,000,000 a year. This does not include a further extension of British effort, agreed a few months ago, by which supplies of materials and foodstuffs from the United Kingdom and the colonies without payment, which had previously been exported against payment in dollars. It seems that the new range of supplies coming in this category include rubber and tea from Ceylon, asbestos, sisal, cocoa, chrome and other essential materials from various colonies. This more recent British offer is mentioned in the White Paper.

The problem of attaching money values to everything done and supplied is very difficult; in some respects it is impossible of solution without a very large staff of experts in this country. The position is less difficult in the United States, where practically all exports for Britain are despatched against large contracts. By comparison British reverse supplies are much more decentralised and difficult to value in terms of money. But it must be clear to all familiar with the facts that the pooling of resources has proceeded well in both countries, and there can be no doubt that such pooling is beneficial to the United States as well as to Britain and the Commonwealth.

The main point to bear in mind is that the British and Americans are partners in this war. In the main, they have the same objective, to free the world from oppression, and they should move forward as comrades, each contributing its utmost to the common cause and without hesitating in an endeavour to compare the value of the efforts made in terms of money.

Mutual aid agreements have been made between this country and the Soviet Union, the French Committee and others, and as the White Paper states, up till May of this year, Britain had delivered 4,690 complete aircraft to Russia, besides spares. But there is nothing in the agreement about the monetary value of reciprocal aid. After all, the nations comprising the United Nations are united towards one end, that of victory, and it should be the duty of each nation to give mutual aid to the utmost of its power, without giving too much attention to the cost in terms of money. It is probable that some sort of balance will be struck when hostilities cease, when it will likely become clear that Britain has raised loans to finance various theatres of war sufficient to hamper her economic recovery; but what do such things really matter if Britain and the United States unite to maintain the shield of freedom in the face of tyranny? Goods, materials and services can be continued almost indefinitely, and while each has its value in terms of cash, however difficult it may be to assess, striking a balance for valiant lives lost in the struggle is more difficult; these are irreplaceable. Such losses can be reduced by continuing mutual aid between the nations without giving too much thought to striking a cash balance.

The fact that goods made of raw materials in short supply owing to war conditions are advertised in "Metallurgia" should not be taken as an indication that they are necessarily available for export.

Planning of the Gas Industry

TOWN'S gas is so important to many industrial undertakings that the proposals for the planning of the gas industry, submitted in a recent report to the Minister of Fuel and Power by the British Gas Federation are of more than ordinary interest. This report is the result of two years' deliberation by a committee appointed to collect and collate relevant information and to consider the steps to be taken to place the gas industry upon such a basis that it can best meet future conditions and make the largest contribution to national and social welfare.

The report recommends that the Ministry of Fuel and Power should appoint a National Fuel Advisory Council to facilitate the elimination of wasteful competition between the fuel industries and the substitution of co-operation in the national interest. Other recommendations are designed to support integrating facilities to expedite and co-ordinate the rational merging, joint operation, or association of gas undertakings or their co-operation in specific activities. Functional integration will be effected by a new British Gas Association, whose Central Council will be the "Parliament" of the Industry. As to geographical integration the report says there is nothing to indicate that the particular needs of the gas industry are such as to require recourse to the "revolutionary expedient" of nationalisation.

Recognising that its full development depends on sustained scientific research, the industry has spent over £500,000 annually in that connection, with substantial results in improved thermal efficiency, cheaper gas, and conservation of coal. Fundamental and long-term research is conducted by the Gas Research Board in association with the Department of Scientific and Industrial Research and is thus co-ordinated with other fuel research. Applied research into the manufacture and utilisation of gas is conducted by the large undertakings and the makers of plant and appliances; Regional Gas Engineering Advisory Boards are securing by local investigation the maximum economy and efficiency in gasworks processes, and Industrial Gas Centres are dealing with problems appertaining to the application of gas to particular heating processes, which are of great importance to the industries concerned.

In future, such matters as applied research, technical development, testing of appliances and materials, and training of personnel will be dealt with by Area Technical Development Centres incorporating the present Industrial Gas Centres co-operatively based on large undertakings.

The gas industry is proceeding with the technical and commercial replanning of its business and is studying such aspects as constant pressure, removal of sulphur, reduction of variations in calorific value and the observance of agreed standards in gas manufacture, distribution and utilisation. While Ministerial sanction will be necessary in many respects, these matters are best resolved by the industry itself, which is so organised as to ensure that its recommendations will be observed.

With regard to manufacture and distribution, co-operation with the Ministry of Fuel and Power and the collieries is necessary to maintain constant quality in the coal processed by each undertaking. Standards of gas and coke manufacture will be observed as the result of the exercise of the agreed authority of the British Gas Association.

Extensive reconstruction of plant and mains will be needed at the end of the war, not merely to restore the *status quo ante*, but to meet the increased consumption which will follow the operation of a national fuel policy. The report favours the inter-connection of an undertaking with adjacent gasworks or coke-oven plants for the supply of gas where works are too small to be economic engineering units, are handicapped by obsolete plant, have difficult access to coal, or for other reasons are relatively uneconomic.

The war has demonstrated the self-reliance, adaptability and reliability of the gas industry in meeting unprecedented conditions, and this report indicates that it is fully alive to the national needs.

Forthcoming Meetings

IRON AND STEEL INSTITUTE.

Nov. 23. Joint Meeting in Scunthorpe with the Lincolnshire Iron and Steel Institute, at 7-30 p.m., in the Modern School, Cole Street. Mr. J. N. Kilby will preside, and Mr. A. Robinson will present the following paper for discussion: "A Review of Basic Open-hearth Practice at an Australian Plant," by Mr. R. L. Knight.

INSTITUTE OF BRITISH FOUNDRYMEN. LONDON BRANCH.

Nov. 24. "Magnesium Foundry Practice," by I. F. Ross. At Slough.

BIRMINGHAM BRANCH.

Nov. 27. "An Atlas of Defects in Non-Ferrous Castings." Non-Ferrous Sub-Committee Report, presented by Mr. F. Hudson. At the James Watt Institute.

SHEFFIELD METALLURGICAL ASSOCIATION.

Dec. 4. Joint Meeting with the Sheffield Society of Engineers and Metallurgists and the Iron and Steel Institute, at the Royal Victoria Station Hotel, at 2-30 p.m. Discussion on the Fourth Report of the Oxygen Sub-Committee.

NORTH-EAST COAST INSTITUTION OF ENGINEERS AND SHIPBUILDERS.

Dec. 10. "The Influence of Production Requirements on the Design of Reciprocating Machinery," by R. E. Strub.

THE INSTITUTION OF ENGINEERS AND SHIPBUILDERS IN SCOTLAND.

Dec. 7. "Gravity Die Castings in Non-Ferrous Alloys," by E. H. A. Carlton.

ELECTRODEPOSITORS' TECHNICAL SOCIETY.

Nov. 29. Annual general meeting at 5-30 p.m., at Northampton Polytechnic. Following the business, an Electrodeposition Brains Trust session will be held, at which the President, Dr. J. R. I. Hepburn, will act as question master. Members are invited to send any question to the Hon. Secretary.

Dec. 7. "The Return of Zinc Plating," by L. Wright, B.Sc., at the James Watt Memorial Institute, Birmingham, at 5 p.m.

The Weldability of Cast Iron

By T. J. Palmer, A.M.Inst.W.

The different types of cast iron are outlined, and certain structural differences are illustrated. A welding process where the heat input per unit volume is relatively low is most suitable for the joining of cast iron, and gas welding methods form the subject of this discussion. Particular attention is given to fusion welding, which remains the most widely used process. It is stated that cracks, hard zones and porosity are troublesome factors, but that such difficulties may be overcome with correct procedure and intelligent operators. The importance of flux and of the use of the rod, which is chemically correct, is stressed. Methods of joining cast iron where deep fusion of base metal does not occur are considered, and it is shown that low temperature methods have certain inherent advantages over fusion welding. The weldability of "burnt" cast iron is discussed, and, to conclude, the present position of the processes under consideration is examined and the outstanding points relating to procedure are reviewed.

THE production of metal shapes by casting fluid iron into moulds is a process which is nearly as old as history, and cast iron has enjoyed an almost unbroken chain of applications since the very early times. It is no less important to-day. In spite of advances made in metallurgical science during the last few decades—which is now reflected in modern steels of 100 tons tensile—simple cast iron, which is but one fraction as strong, still finds a useful place. The development of alloy cast iron is significant, and is indicative of future trends as increased knowledge of the metallurgy of this aged metal is obtained. Ordinary cast iron, by virtue of its low cost, is still used extensively, and whilst this material, which is so inherently lacking in ductility, is employed there will always be scope for "welding." Welding as applied to cast iron is the most satisfactory means of repairing fractured parts, and, with correct procedure, uniformly good results can be obtained. However, among the many different metals now being welded, cast iron is probably one of those which is least fully understood, and often there appears to be some uncertainty on the part of both engineer and welder as to which process or technique is most suitable. A discussion of the subject is therefore in order. Malleable cast iron, a product of the annealing furnace, is not included in the present discussion.

Fortunately, from a scientific point of view, the welding of cast iron has been probed fairly deeply, especially by those concerns which undertake to supply filler rods for the purpose, and it is due, largely to their combined efforts that cast-iron welding has become commonplace to-day. This does not mean that selective choice of filler material eliminates the difficulties which have always been associated with the process—or that cast-iron welding has been made foolproof. The pitfalls are many, and the subject calls for a knowledge of the properties of this super-brittle metal, as well as for skill and perseverance on the part of the welder.

Types of Cast Iron

The study of cast iron is a large subject, but the principal grades at present used on any great scale fall directly into two classes, white and grey, the classification having arisen from the appearance of the metal after fracture. The chief difference between the two is that with white cast iron the major part of the total carbon present exists as Fe_3C, which, in massive form, dominates the microstructure (Fig. 1); with grey cast iron, graphite, in flake form, occupies this position (Fig. 2). A brief description of the two grades is given.

White Cast Iron.—The microstructure of a typical white iron is shown in Fig. 1. It consists, essentially, of thick walled grains of Fe_3C enclosing the eutectoid pearlite, and the phosphide eutectic. In view of the hard and unworkable nature of white iron its applications are limited; the larger part of this material is absorbed by annealing furnaces, which convert it to malleable iron. Commercial applications of white cast iron are shown in the production of crusher plates, brake segments, metal rolls, draw dies, etc., and for certain other purposes where resistance to wear and abrasion is a primary requirement. White cast iron can be welded if suitable precautions are observed.

Grey Cast Iron.—Grey cast iron is used to a far greater extent than the type just described, and it forms the base of practically all engineering iron castings. Three types predominate:—

Ferritic Cast Iron, consisting of plates, or flakes of graphite set in a matrix composed mainly of ferrite (Fig. 3).

Ferritic-pearlitic Cast Iron, where the matrix contains both ferrite and a fair proportion of pearlite (Fig. 4).

Pearlitic Cast Iron, containing flake graphite in a ground-mass of pearlite (Fig. 5).

Any phosphorus present forms a separate component. This can be seen faintly in most of the accompanying photomicrographs; in Fig. 6 its distribution is more clearly marked as a result of prolonged etching. Under certain conditions of cooling free carbide areas will break up the normally "grey" structure, and this is illustrated in Fig. 7. The microstructure of grey cast iron can be modified very considerably by cooling speed, composition of iron, and also by any impurities present. The welding of grey iron is practiced widely.

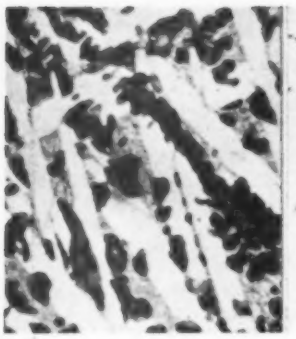


Fig. 1.—White cast iron showing carbide.

Fig. 2.—Graphite flakes in cast iron.

Fig. 3.—Ferritic cast iron (graphite and ferrite with some pearlite).

Fig. 4.—Ferritic-pearlitic cast irons (the mainly pearlitic matric is broken by graphite with small ferrite areas).

Alloy Cast Irons

In addition to carbon and silicon, which are always present—both of which exert a fundamentally important influence over the properties of the iron—special elements are frequently added with the object of producing "alloy" cast irons where increased hardness, toughness, corrosion or heat resistance is desired. Nickel, chromium, molybdenum, vanadium, manganese and copper are the most important of these elements. Their individual effects, depending upon the amount added, may be to stabilise the carbide, and to encourage its precipitation during cooling, as is the case with chromium, vanadium and others, or to favour and control graphitisation—a particular function of nickel. The effect of alloy content upon the metal matrix is no less considerable; it can be refined or hardened by the substitution of fine pearlite, sorbite or martensite for relatively coarse pearlite, or, under favourable conditions, it may be rendered austenitic. In addition, alloying elements may be both ferrite forming, and ferrite strengthening. Many alloyed irons are amenable to, and are improved by, heat-treatment. Most alloy cast irons can be welded.

Welding

Cast iron is welded by both the electric arc and gas welding processes. The arc welding of cast iron has, of late, received notable attention, and there are certain instances where it offers advantages. However, there is little doubt that in this case gas welding is more suitable. Particularly is this so if one is dealing with castings in which there are extreme variations in metal section, and where, as a result of this, the tendency for cracking during welding, or for dangerous residual stresses to be retained, is much greater. Gas welding is the process which is most widely employed, and it will therefore receive preference in this discussion. The methods of joining cast iron by this process can be placed under three main headings—"Fusion-welding," "Low-temperature welding," and "Bronze-welding." They are equally important at the present time and will be discussed individually.

Fusion Welding

Fusion welding, which involves liquifying the base-metal edges and the addition of a filler rod of approximately similar melting point, composition and colour, is the method which has been employed since the very early days of welding. The bond is homogeneous (Fig. 8), and has colour characteristics comparing very favourably with those of the parent metal; the latter is an important consideration.

Certain difficulties have always adhered to the fusion welding of cast iron—difficulties which are common to all types. They are—the prevention of cracks; the formation of a machinable deposit; the avoidance of porosity. To overcome these difficulties requires consideration, skill and care.

The Prevention of Cracks

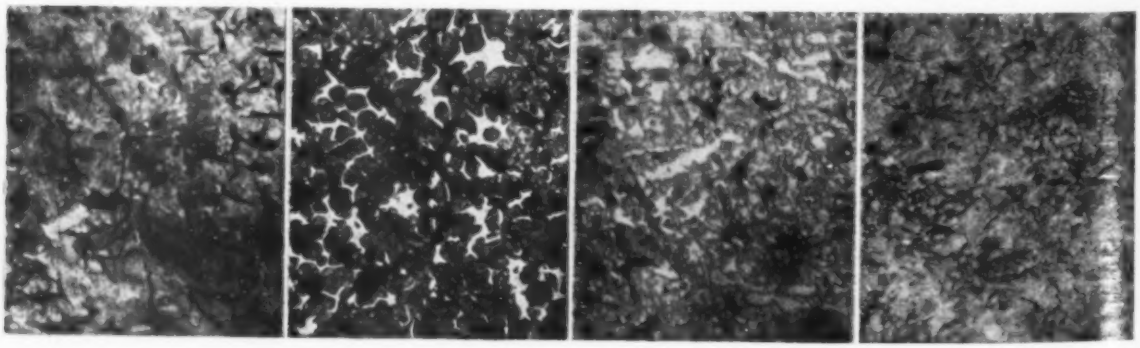
The fact that most cast irons have no "give" does not require emphasis; it has been brought home forcibly to both engineer and welder too often. Partly

Fig. 5.—Pearlitic cast iron (graphite and pearlite).

Fig. 6.—Distribution of phosphorus in cast iron (white areas).

Fig. 7.—Grey cast iron fairly quickly cooled, showing free carbide areas.

Fig. 8.—Micrograph of fusion weld at junction of weld and parent metal.



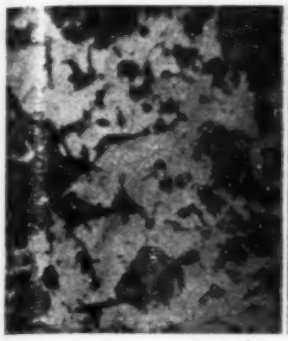


Fig. 9.—Close up of hard spot in cast-iron fusion weld (carbide areas shown light).

Fig. 10.—Sand-cast iron (graphitic structure).

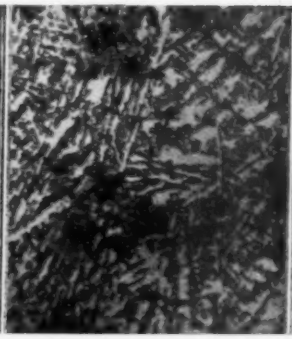


Fig. 11.—The same iron as Fig. 10, but centrifugally (chill) cast. Structure contains free carbide.

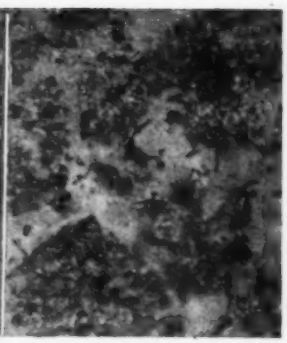


Fig. 14.—Structure of a modern high-silicon welding rod as used in fusion welding.

by reason of undisturbed cooling from liquid most "cast" metals are left in a more or less brittle condition. This property is of paramount importance in welding. In making a weld parts are heated locally, and in fusion welding they are heated to their melting point, which in this case is about 1,200° C. Stress gradients will therefore be set up, due to differential expansion movements of the part. The reverse, complicated by shrinkage of the joint, takes place upon cooling, and the inability of cast iron to accommodate these volume changes frequently results in the formation of cracks. Slow heating and retarded cooling are therefore of fundamental importance in the fusion welding of cast iron. In this manner temperature and stress variations are minimised during heating, and those developed from, or incidental to, the welding operation proper are distributed more freely upon cooling. Preheating and post annealing serve this purpose, and the methods employed are varied. Furnace heating is, of course, the best, but improvisatory methods are often used; these are generally quite satisfactory. The points to be stressed are that the heating range (650°–750° C.) should be reached gradually, the weld made rapidly, and the part cooled slowly. Not all cast-iron parts require *full* preheating, and the procedure is less exacting where castings are not complicated, but in the absence of preheat the danger of cracking is increased when the part is thin; where wide variations in section are a character of the casting; and where the presence of struts or ties offer resistance to normal thermal movement.

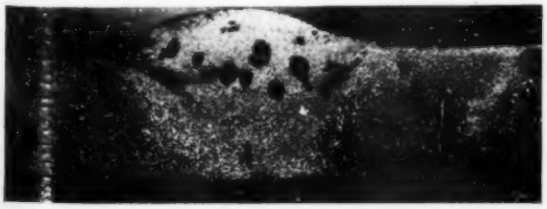


Fig. 12.—Fusion weld in cast iron. Blowholes resulting from incorrect working procedure.

With intricate castings, where long seams are involved, or where a number of runs are necessitated, the welding time becomes prolonged. To avoid wide temperature variations in the part is is frequently necessary, under such circumstances, to interrupt welding and to apply further preheat. Intermediate heatings will flatten temperature gradients, which are formed if the welding time is long. Failure to do so may mean the accumulation of stresses within the casting, sufficiently high in magnitude to cause rupture at some distance from the point of welding. Additional heating is frequently required, even if the initial preheat involved the uniform heating of the entire article. Whilst there is no point, of course, in wasting time and gases for the reheating of castings which do not justify such treatment, it is certainly not worth the risk of inviting fracture—simply by failing to apply it when required.

The Formation of a Machinable Deposit

A most vexing problem when surfacing a cast-iron part which has been welded is to encounter a hard deposit, or hard spots, in what is otherwise a fairly soft deposit. The fear of this has often led to substituting a new casting for a fractured one which could have been both easily and satisfactorily repaired by welding. Hard spots, or zones, need not occur if proper precautions are observed, and it is far better to avoid their formation than to try to break down any hardness later. Hardness in cast-iron welds is, in most cases, a result of carbide formation (Fig. 9) and the manner in which it is formed

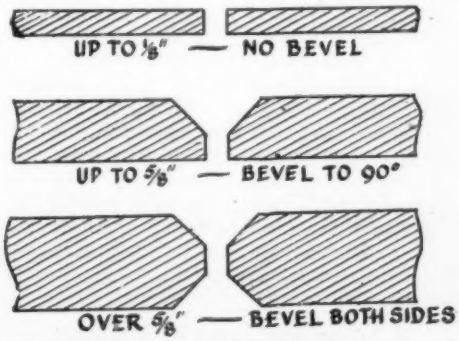


Fig. 13.—Method of preparing metal edges for cast-iron fusion welding.

may be complex, for it is bound up with the mechanism of crystallisation as the metal cools from the liquid phase.

Temporarily neglecting the presence of other elements, the cast-iron weld melt consists of a solution of iron

and carbon at temperatures above the liquidus. Whether the latter element is dissolved completely, or if small solid particles remain, probably to function as nuclei for eventual solidification, is immaterial for purposes of this discussion. In any case this may not be so in welding because of the high degree of superheat frequently attained by the melt. Upon cooling carbon is precipitated, due to the fact that its solubility in iron decreases with a falling in temperature; this is in accordance with the iron-carbon equilibrium diagram. In the early stages of solidification carbon deposition probably takes the form of Fe_3C (carbide of iron) which, if conditions are favourable, later breaks down to graphite. Thus, bearing in mind the modifying effects of added elements and impurities, cast irons of normal composition are predominantly graphitic when cooled slowly. In this case they are soft and easily machined; or, under conditions of more rapid cooling, they may contain much free carbide, in which case it will be difficult, if not impossible, to machine them. Figs. 10 and 11 serve to illustrate this point.

In making a weld the operator must produce a joint which is machinable, and in order to do this the factors conducive to chilling must be mitigated. Deposition of weld metal upon surfaces which have been preheated insufficiently will, for example, invite the presence of free carbide in the finally cooled product; the discrepancy in temperature between added, and parent metal causes chilling of the former as its heat is rapidly abstracted by the much cooler casting mass. It is important then, particularly when preheating is locally applied, that the temperature to which heating is carried, and the area over which it is spread, be such that rapid flow of heat from deposit to base metal cannot occur. In practice, this means merely that before welding can commence the regions around the proposed joint are preheated adequately.

Direct chilling of the melt, or portions of same, will also produce hard spots in welding. For instance, the addition of a rod of very large section, or "plunging" a dead-cold rod into the molten pool can create such an effect; both are points which should be watched, and care taken to guard against them. Unsuitable filler rods are yet another cause of hardness.

The loss of silicon in welding is particularly bad, since its presence in cast iron is most valuable in opposing carbide stabilisation. Its loss through oxidation or volatilisation cannot be entirely prevented in the welding operation, unfortunately; fillers rods contain it in excess to adjust this loss. To minimise loss of this element during welding, the joint should be made as rapidly as is practicable. Avoid overheating, and the use of an oxidising flame; both will cause its rapid disappearance from the melt. To prevent the overheating of cast iron during welding, the flame must not be held too close to the work. It is fatal for the point of the flame to contact the melt; this will invariably produce hard spots.

The Avoidance of Porosity

The difficulties accompanying efforts to produce a "cast" with complete freedom from blowholes are appreciated by both founder and welder, to whom they are a constant concern and a common curse. Welding is really a casting process, and though, in this case, the amount of liquid metal involved is small, the operator is confronted by conditions similar to those which prevail in the foundry. As in the foundry, the elimination of blowholes presents a major problem. Blowholes in welding are attributed to the presence of gas, formed either from reactions occurring in the melt, or by direct absorption from the air or flame. If these gases do not surface before solidification, blowholes will almost certainly be formed. The latter take the shape of both round and elongated pores, which can range in size from large cavities down to pin-holes, and an incorrectly made fusion weld may evidence a considerable amount of porosity (Fig. 12). An analysis of the factors involved in welding indicates that certain critical conditions favour the formation of blowholes, and the following are mentioned as being influential:—

(1) Oxide, rust and foreign matter present on the metal surface as a result of imperfect preparation of work.

(2) Sand specks and other non-metallic impurities of a refractory nature in the parent metal.

(3) Dross, etc., on the welding rod.

(4) Insufficient flux.

(5) Overheating of the weld melt.

(6) Oxidation of filler metal in its passage to the melt.

(7) Too low a working temperature.

(8) Excessive agitation of the melt.

(9) Direct admission of oxygen into the melt, either from an excess in the flame, or from contacting the metal fluid with the flame cone.

A proper study of the above is necessary if best results are to be obtained. Statements to the effect that blowholes are entirely avoidable in welding might be open to discussion, but under properly controlled working conditions, and with reasonable operating skill, their formation can be limited to small, if not negligible, proportions.

Procedure of Fusion Welding

Parts are first degreased and then prepared in a normal manner according to thickness of section; with material above $\frac{1}{8}$-in. thick this entails filing, grinding or chipping of the edges to form bevelled joints, as illustrated in Fig. 13. Oxygen cutting is neither a convenient nor satisfactory method of effecting weld preparation in cast iron. It is desirable to expose fairly clean metal where filler material is to be deposited, in order that any surface sand, etc., does not become a possible source of trouble. Protection is afforded to highly polished surfaces by the use of an approproate preparation.

A flux containing $NaCO_3$ and $NaHCO_3$ is generally employed. The fusion temperature of oxide of iron is about 1,275° C.; this is higher than the melting range of most cast irons. Solid particles of oxide may therefore be encountered which, unless eliminated, will not only deprive the metal of its fluidity, but will contaminate the finished joint. A good flux precludes this effect. It is applied as an adherent tuft from the heated welding rod end, or/and by sprinkling over the heated work surface. Welds made without flux are unsatisfactory. At the same time an excess can be bad; this serves no useful purpose and forms a hard slag film which is difficult to remove.

Attention should be given, in setting up the work, to ensure free access during welding, and a proper sequence of weld runs is determined where complexities of casting design demand it. Care must be taken against

(*Continued on page 25.*)

Jigs and Tools for Anodising

G. O. Taylor, M.S.E., A.I.P.E.

In a previous article by the author consideration was given to the design of jigs and tools for the handling of light metal articles during the process of anodising. In this complementary article designs for further jigs and tools are described that can with advantage be added to the equipment of an anodic oxidation plant to speed-up production and reduce labour and material costs.*

THE greatly extended usage of the anodic oxidation process for the treatment of cast and wrought aluminium alloy products, particularly in the aircraft industry, has led to considerable ingenuity in the design of jigs and tools to enable plants formerly working more or less on a "jobbing" basis to swing over to mass production methods.

In many plants devices to enable the handling of a large number of parts by a single operative were previously almost non-existent, and even some months after the commencement of the war the anodising department of a large aircraft factory visited by the writer was using hundredweights of pure aluminium 18 s.w.g. wire each week for the purpose of effecting individual contact to standardised components that were ideally adaptable to processing on jigs. Individual wiring practice of this kind is not only expensive and wasteful, since the thin wire can be used efficiently only once, or twice if it is carefully straightened and the anodic film stripped off in hot caustic soda solution; but it is also unnecessarily extravagant from the aspect of the labour required to handle a large number of parts.

Additionally, it is rarely possible with individual wiring to secure the full potential output from a plant of given size, since the processing vat cannot be loaded with the maximum number of parts within the capacity of the generating equipment available, owing to the necessity for the wider spacing of the parts to be anodised to avoid "burning" due to components contacting each other, or the sides of the vat, if set in motion by the agitation of the electrolyte.

Obviously, under such conditions it is also difficult to make full use of the depth of the process tank.

If consideration is given to the design in suitable aluminium alloys, of jigs and tools made from extruded sections, sheets, tubes or round rod of $\frac{1}{8}$ in. diameter upwards, there is little or no occasion to use thin wire for contacting purposes. Furthermore, such appliances can be used over and over again in the treatment of large numbers of parts before it is finally necessary to scrap them, thus effecting considerable economies not only in labour and aluminium, but also in production costs, by reason of the increased output rendered possible by the larger batches that can safely be loaded in the vat. Again, such jigs can often be fabricated from odd pieces of material that would otherwise be consigned to salvage.

In the designing of jigs and tools for anodising there are several fundamental principles that must be borne in mind:—

(*a*) The electrical contacts made to the part to be anodised should be as small as possible, but large enough to pass into the part the required amperage without causing overheating and burning of the metal at the point of contact. This factor is, in general, more important in the case of the sulphuric acid process, where greater current densities prevail, than in the chromic acid process.

(*b*) The part must be firmly gripped to prevent movement of the contact points by agitation during treatment. The oxide film formed on the metal is electrically insulating, and if the metal-to-metal contact is lost current will cease to flow into the part being anodised and it will not acquire a normal film thickness.

(*c*) A welded jig construction should be avoided, as welds tend to deteriorate on continued exposure to the electrolyte.

(*d*) Jigs designed to hold a multiplicity of parts should be so constructed that anode-cathode distance is preserved approximately the same from each part.

(*e*) No metals other than suitable aluminium alloys should be used in any part of a jig that is immersed in the electrolyte, but the use of plastics or plastic compounds is permissible, and advantageous in some cases. Metals other than aluminium can, where necessary, be used in jig parts not immersed in the electrolyte.

(*f*) Jigs for components that are hollow, or have recesses, should be so designed that the openings in the components face upwards to avoid the formation of gas or air pockets which result in an untreated area on the metal.

(*g*) Special care must be taken in the design of jigs for handling castings, as, in general, cast alloys consume considerably more current than do wrought alloys, particularly at the commencement of treatment. Jigs must be of stouter construction to pass the higher amperages involved.

(*h*) The best alloys to use for tool construction are those that consume the lowest amperage per square foot, so that maximum current is consumed by the articles under treatment. Suitable materials are Duralumin, pure aluminium, aluminium-magnesium, or aluminium-manganese alloys.

(*i*) Wherever possible the jig should be designed to allow contact points to be cleaned after anodising by use of a smooth file or emery-paper. This enables the jigs to be put into use without need for stripping off the anodic film in caustic soda or acid etches, and greatly extends their useful life. It also enables the possibility of treating all parts of the jig other than contact areas with suitable "stop-off" materials, such as bakelite-type varnishes, synthetics, lacquers, or stoving enamels, which should be applied after the jig has passed through the anodising vat once and has acquired a good anodic film. This treatment further extends the useful life of the tool.

In the previous article four designs of jigs for handling small and medium-sized parts were described. The features of these tools will be briefly recapitulated to link up with the additional designs to be illustrated later.

* METALLURGIA, July, 1940.

Bent Rod Jigs

These types of tools are made by bending $\frac{1}{8}$ in., $\frac{3}{16}$ in., or $\frac{1}{4}$ in. diameter aluminium alloy rod of "springy" temper into shapes appropriate for gripping the articles to be treated. In general, they are of four basic types: (*a*) Multiple outside grip. (*b*) Multiple inside grip. (*c*) Single outside grip. (*d*) Single inside grip.

The multiple outside grip is made by bending a length of rod to a serpentine form of dimensions that will allow standardised small parts to be gripped on opposite outside edges. Nicks are filed in the jig at points suitably disposed inside the bends, to enable articles to be sprung in and gripped firmly in alternate left and right bends. This type of jig will handle small parts without recesses or holes.

The multiple inside grip is similar in shape, but is adapted for holding standardised parts containing a hole or recess. In this case the articles are pushed on to opposite alternate left and right bends, each U-shaped portion of the jig being arranged to spring outwards inside the hole or recess to effect tight contact.

Single inside or outside grips are similar in principle to the multiple type, but are generally made in the stouter diameter rod for holding single larger articles. Inside grips are pushed into a hole or recess in the component and are tensioned to spring outwards, while outside grips are sprung over opposite points on the outside of the part to be anodised.

Compression Rod Jigs

This type of jig consists basically of a length of round rod provided with a stop at one end, and will accommodate either a number of identical parts or a miscellaneous assortment of parts made from one alloy; but in all cases it is necessary for the components to have disposed at some point a hole somewhat greater in diameter than the rod to enable them to be threaded on it. As the parts are threaded on the rod they are spaced from each other by pieces of tubing or half-round rod drilled $\frac{1}{16}$ in. larger in diameter. Above the spacing-piece resting on the uppermost article a length of tube sufficient to project well above the electrolyte level is placed, then the miscellany of parts threaded on the rod are forced into tight contact by the action of a steel compression spring dropped over the rod to rest on the tube, compressed and retained in compression by a suitable clamp.

Finger Grip Jigs

This is one of the most useful tools, and consists of a main conducting member of $\frac{3}{4}$ in. square rod, which accommodates 10 in. lengths of $\frac{1}{8}$ in. diameter heat-treated Duralumin-type alloy wire passed through holes drilled in the rod and then peened into position to project equally from either side. Alternate wires are passed through the main member at right-angles, so that "fingers" project from all four sides.

A large number of small articles can be loaded on a single jig of this type, the projecting wires acting as spring fingers to effect contact.

Bolt and Bar Jig

This jig is made by riveting half-round extruded section on both sides of an extruded flat section and then drilling at regular intervals through the composite member so formed. A miscellany of parts can be processed on a tool of this type with the aid of light alloy nuts and bolts and some odd lengths of half-round section drilled to take the bolts, which must also pass through the hole in the main member. Parts to be treated are held in position against the half-round face of the main member by the nuts and bolts in conjunction with a piece of half-round section lodged across the part, or across any suitably disposed hole that may be present in the component. Multiple point contact is effected, and the use of half-round section ensures small contact areas.

The construction and uses of the jigs described are fully discussed and illustrated in the July, 1940, issue, but have been referred to in order to join up the additional designs detailed in the present issue.

Spring Jigs (Fig. 1)

A very useful device for handling a multiplicity of small standardised parts can easily and quickly be made by winding a length of $\frac{1}{8}$ in. or $\frac{3}{16}$ in. diameter rod in a "springy" temper alloy around a cylindrical former 1 in. to $1\frac{1}{2}$ in. in diameter to produce a tightly coiled spring about 2 ft. long, sufficient rod being left straight at one end to engage in a suitable clamping device on the bus-bar.

Small parts are contacted by prising open adjacent coils, inserting a flat surface of the article and then allowing the coils to spring back into position. Alternate left and right sides of the spring can be used in this manner to engage a large number of parts.

By the use of two such springs, lengths of drawn

Fig. 1.—Supporting small parts between adjacent coils of a spring jig.

section or the thinner extruded sections can be handled with speed and ease, flat surfaces of the sections being inserted between coils of the springs which are disposed at the ends of the lengths.

After treatment of the first batch of parts and after each succeeding batch, it is necessary to remove the anodic film from the jig by etching in hot caustic soda solution, followed by swilling and neutralising in nitric acid solution, to present fresh contact surfaces. A large number of parts can be anodised before the spring is weakened too much to allow further use.

Expanded Sheet Jigs

Pure aluminium expanded sheet, which is made in a range of sizes of mesh openings, is a useful material, allowing scope for the design of some ingenious methods of contacting parts to be anodised. A length of the sheet

clamped at both ends between two flat extruded strips (suitably extended to contact the bus-bar) can be employed in a variety of ways.

Small components in the stronger and harder alloys can be wedged in the numerous openings available, and are held in firm contact by the slight " bite " made into the softer pure metal.

The multiplicity of openings also provides an excellent means of handling large pressings awkward to jig up by other means; since the pressings can be held tight against the mesh by short lengths of section bedding on any suitable surface and held firmly in position by light alloy nuts and bolts passed through the mesh and tightened on to a second piece of section on the opposite side.

Components such as castings and forgings that are fairly heavy in relation to size and which will pass partly through the mesh but have protuberances which prevent their falling through, can be left depending and will effect contact by virtue of weight.

In this case it is preferable for the expanded sheet to be arranged in a horizontal position in a suitable framework of extruded section. With jigs of this type it is necessary to remove the anodic film with hot caustic soda solution before each load of parts is anodised.

Spring Clamps

A form of spring-clamp that can easily be made and offers a facile means of handling sheet components drilled for ultimate riveting is illustrated in Fig. 2. It consists of two pieces of " springy "-temper sheet riveted to an extruded flat section or to a strip of thick sheet extended to contact the bus-bar. The spring temper sheet which forms the jaws of the clamp is suitably bent, as shown, to spring together again if prised apart and released.

One of the jaws of the clamp is drilled to take as a push-fit a small length of heat-treated Duralumin rod tapered as shown and held in position by peening with a centre-punch. On the opposite side the clamp is drilled to take as a push-fit a short length of tube somewhat greater in diameter than the holes in the component, and this also is held in place by peening.

The action of the clamp in effecting a rigid contact is as shown, and two or three of such clamps can be disposed along a drilled sheet. All that it is necessary to do between batches to restore contact on the tapered point is to remove the anodic film by rubbing with emery cloth.

Tools for Bulk Anodising

The possibility of treating small articles, such as rivets, washers, screws, nuts, pressings, castings and stampings in bulk without the necessity for ensuring individual contact to each part has been explored to some extent. Suitable types of tools are now in use, but this branch of jig design has not yet been exploited so fully as it might be.

The principle of bulk anodising is that standardised small parts are loaded in suitable containers and shaken into minimum space by tapping the container with a mallet. With the exception of flat surfaced articles, such as washers (which demand special attention), small parts packed in such manner contact each other only at portions of their surfaces and there is sufficient space between them for electrolyte to percolate and for the " gas " evolved during anodising to escape.

Contact to the mass may be effected via the sides and bottom of the container in conjunction with a pressure plate contacting the uppermost layer, by means of top and bottom plates only, or by contact with the container alone. With good tool design an average of 99% or more of small parts loaded in a container will be completely anodised, and many of the remaining 1% or less

Fig. 2.—The design and application of one form of spring clamp.

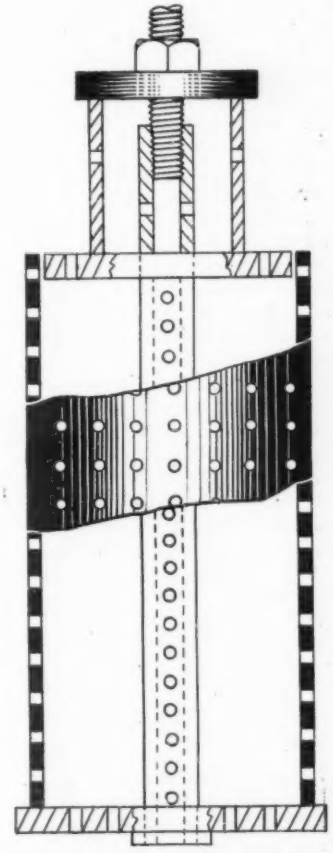


Fig. 3.—An anodising container for the bulk treatment of rivets, nuts, eyelets and similar small articles.

partially anodised, only very few failing to acquire any film at all.

A very efficient tool for the anodic treatment of rivets, nuts and bolts, eyelets and similar small articles, is illustrated in Fig. 3. The container, which may be 4 in. to 6 in. in diameter and 12 in. to 18 in. long, need not be of light alloy, the life of the tool being considerably prolonged by making this part in a plastic material, such as a suitable form of bakelite. " Tufnol " is ideal for the purpose.

This tube is liberally perforated with $\frac{1}{8}$ in. or $\frac{3}{16}$ in. diameter holes, and is fitted with a central tube $\frac{3}{4}$ in. to 1 in. in diameter in a suitable light alloy. The central tube is also perforated and fitted at the bottom with a thick, stepped aluminium alloy plate, as shown. Screwed into the tube is a threaded steel rod terminating in a removable hook for contacting the bus-bar. Small parts are loaded in the container and tapped into minimum

bulk and a top plate of light alloy is dropped over the tube and forced into contact with the uppermost layer by means of a nut engaging on the threaded steel rod, applying pressure through a spacing tube.

Such a tool will handle thousands of small parts, such as, for instance, rivets in one operation, and is economical in upkeep since the container is not affected by the electrolyte, and it is necessary to remove the anodic film only from the contacting surfaces of the top and bottom plates between batches.

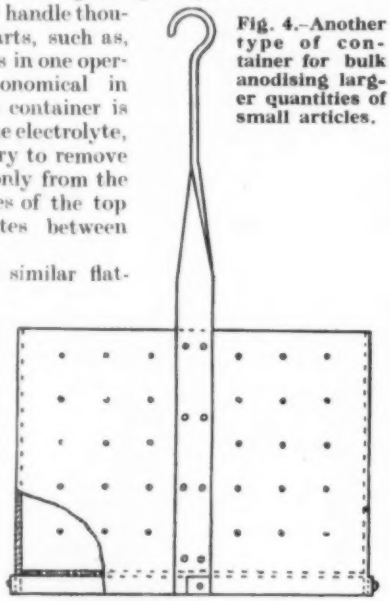


Fig. 4.–Another type of container for bulk anodising larger quantities of small articles.

FIG. 4

Washers and similar flat-surfaced articles cannot be anodised successfully alone, due to the overlapping effect; but excellent results can be obtained by mixing them with rivets, and ultimately effecting separation by screening on a suitable mesh through which the rivets will fall while the washers are retained.

The problem of drying small parts after anodising and washing can best be overcome by rotating them in a small centrifuge for a few minutes.

Another type of bulk anodising container that will handle up to 56 lb. or more of small articles in one operation, but which is somewhat less efficient than the type illustrated in Fig. 3, is shown in Fig. 4. The perforated container in this case is 12 in. to 18 in. in diameter and depth, and is made from $\frac{1}{8}$ in. to $\frac{1}{4}$ in. thick pure aluminium for preference, with a riveted strap-joint.

Two pieces of flat extruded section housed into each other at the centre to form a level cross are loosed into the container edge-on to ensure the maximum rigidity, and the ends are bent over and riveted as shown. A perforated circular blank of thick pure aluminium sheet rests on the extruded sections to form the bottom of the container.

Two pieces of flat extruded section (one of which may be the strap covering the joint in the container) are riveted to the container and provided with a twist to enable the bus-bar to be contacted by the looped-over ends.

No top plate is provided, the weight of the mass of parts in the container being relied on to provide contact. It is necessary to strip the anodic film from the entire assembly between batches of work.

Jigs of the types described are easy to fabricate, and will repay their cost many times over in man-hours saved in the effecting of efficient contact to articles to be anodised. The factor of safety provided against parts contacting each other or the sides of the anodic tank during processing enables the full capacity of the generating equipment to be utilised more often than where individual wiring is practised, with consequent greater turn-over from the available tank capacity.

The Constitution of Magnesium-Manganese-Zinc-Aluminium Alloys

THE work described in a recent paper by A. T. Little, Dr. G. V. Raynor, and Dr. W. Hume-Rothery* is a continuation of a systematic investigation of the structure of quaternary alloys of aluminium with magnesium, manganese and zinc, in the range magnesium 0–5, manganese 0–2, and zinc 0–8%. The present paper is concerned with the constitution of this group of alloys at 500° and 400° C., the former temperature being very slightly above that at which the whole group of alloys is completely solid. The alloys used were prepared from the same metals as those described in the paper dealing with the liquidus.[1]

At 500° C. the solubility of manganese in aluminium is greatly reduced by the presence of magnesium and/or zinc. In the ternary system magnesium-zinc-aluminium, nearly all of the present range of alloys are of the simple α-primary solid-solution type at 500° C., except for a few alloys with the highest contents of magnesium and zinc, which are just within the (α + liquid) area. In the quaternary system, the boundaries between the α (α + $MnAl_6$), and (α + $MnAl_6$ + liq.) fields have been established at 500° C.

At 400° C. the solubility of manganese in the alloys is negligible, and all alloys examined contained the compound $MnAl_6$. In the ternary system magnesium-zinc-aluminium, the alloys studied lie either in the α or the (α + T) fields, where the symbol T is used to denote the intermediate ternary phase based on the composition $Al_2Mg_3Zn_3$. The solubility curve has a characteristic form, which is discussed theoretically. In the quaternary alloys the extent of the α solid solution is negligible, and the alloys are of the (α + $MnAl_6$) or (α + T + $MnAl_6$) types; methods are described for calculating the boundaries between the two fields.

* *Jour. Inst. Metals*, 1943, **69** (October), 423–440.

1 E. Butchers, G. V. Raynor, and W. Hume-Rothery. *Jour. Inst. Metals*, 1943, **69**, 209.

The Lincoln "Weldirectory"

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WELDING has expanded rapidly to meet war-time requirements, and is now being used in many ways, from the construction of ships to the maintenance of earth-moving equipment to clear aerodromes. In order that the operator may have procedures, properties and correct settings for quick reference on the job, and in order to keep within the paper quota, the above publication has been revised and issued in pocket form. Much of the information is condensed, but the salient features are given in this compact booklet of 32 pages. It is available free on request from Lincoln Electric Co., Ltd., Welwyn Garden City, Herts.

Difficulties Encountered in the Heat-Treatment of Drop Forgings

By Bernard Thomas, F.Inst.P., F.Inst.F.

Heat-treatment begins with the heating of the steel preparatory to forging, and concludes with the final cooling, followed by the last heat-treatment operation. Many of the difficulties encountered by the heat-treater, however, may be due to lack of control of heating in the mill or in the forge shop during shaping, but in this, his Presidential Address to the Staffordshire Iron and Steel Institute, Mr. Thomas assumes the steel, of which the drop forgings are composed, is in proper condition, and confines his remarks to heat-treatment difficulties.

FEW branches of industry can equal the developments made by that of the drop forgings during the war period, as most of the mechanical requirements for the pursuit of hostilities necessitate the products of such in their construction. As probably 95% of modern drop-forgings are heat-treated before entering service, the process is practically synonymous of the product. Unfortunately, however, as with most processes, there arise occasional difficulties, and the object of these remarks is to discuss some encountered in the experience of the author.

There are in total roughly twenty different types of steel, alloy and otherwise, in popular use, for the production of drop-forgings and stampings, despite the steady pressure being brought to bear on the utilisation of substitutes of lower alloying element content. For the most part, all these steels are covered in their governing specification by standard temperatures recommended for their treatment, but it must be appreciated that when any given temperature may be named, it rests afterwards with the user as regards modifications which may be necessary by reason of ruling thickness of the forging.

At this stage it may be advisable to define the expression "ruling thickness." Most drop-forgings are naturally irregular in shape, with their lugs and bosses, etc., and the ruling thickness is usually taken by most as being the general, or average thickness prevailing. Another school of thought suggests that the ruling thickness is the diameter of the largest sphere it is possible to machine from any part of the component concerned.

The tempering direction is usually generous in its breadth of expression, as may be gathered from the following in regard to one of the most popular steels employed :—

"They (the forgings) shall then be tempered by heating to a suitable temperature of not less than 560° C. nor more than 660° C. and quenching in water or oil, at the option of the manufacturer, to give the tensile, Izod and hardness values specified. . . ."

With such explicit instructions, therefore, the heat-treater undertakes his task !

Normalising

Dealing first with the simpler forms of treatment—i.e., normalising—there is probably more En 8 (3S6) material used in this category than all other steels combined. As a 0·30–0·45% carbon steel with a required ultimate stress of 35–45 tons per sq. in., the chief difficulty lies frequently in the obtaining of minimum Izod requirement of 10 ft.-lb. Failures are frequent, accompanied by very coarse fractures, and many of these may be due to the use of a fixed treatment temperature despite the potential variance of 15 points of carbon and 100 points manganese. Obviously, the range advised of 830°–860° C. cannot meet adequately the potential variance of analysis, and enlarged grain size will result if a temperature too high be employed.

Beyond the foregoing, the material may be suffering from overheating or burning during production. If suffering in the latter respect there is no remedy, but the former difficulty may be overcome by suitable treatment according to the degree affected.

It happens occasionally that too little work has been put upon the drop-forging in its production, with the result that the banded structure induced by rolling still obtains. Such cases invariably yield extremely poor physical figures and the only remedy is by conversion of the whole of the pearlitic structure into the solid solution of austenite, and after a reasonable soaking period calculated to allow of the free ferrite completely absorbing and so diluting the eutectoid areas, cooling below 600° C. and reheating for normalising in the ordinary manner. Izod values of below 10 ft.-lb. have been raised to above 90 ft.-lb. by such treatment.

To illustrate further the importance of work-effect, a case is cited of a lug-forging where the shank was half the ruling thickness of the head, or boss, portion. To produce it, the stamping received twice the work on the shank as compared with the head, and a number of experimentation tests showed that the Izod values obtained in the shank were always on the average 50% higher than in the head, and obviously the heat-treatment was of necessity the same in each case.

Where failure of tests occurs in the absence of a banded structure, it sometimes brings in the question of correct treatment temperature. Enlargement of grain size will result in a low Izod value, and to avoid such it may be expedient to employ the Ac2 temperature, which exists for steels up to 0·45% carbon, at almost a stable point, instead of the Ac3 temperature, as although the latter is necessary for complete solution of the eutectoid, grain refinement would appear complete at the former, where magnetism is lost and certainly better physical results are invariably provided.

Other specifications which call for normalising, such as S71, S67, S90, etc., do not as a rule give rise to much difficulty, but the treater should endeavour to be *au fait* with his specifications, as otherwise he may be faced with an attempt to "normalise" an En 36 or 2S28 because his customer has asked for it. Unfortunately, in heat-treating at any rate, the customer is not always right, and it is necessary that his interests should be safeguarded adequately.

Annealing

The annealing of 15 years ago, where a healthy soak at 850° C. yielded wonderful results, has gone for ever with the entry of alloy steels. Annealing, like hardening, requires close control of temperature, as may be determined when it is expressed that by air-cooling certain steels instead of furnace cooling, tensiles up to 110 tons per sq. in. are obtainable.

Many nickel-chromiums and nickel-chromium-molybdenum types, such as S82, 2S28, En 36, etc., harden-up on cooling after forging. In this condition they are unmachinable, and being mostly gear steels, which require machining prior to carburising, annealing becomes necessary accordingly. Even specifications for subsequent full heat-treatment are now sometimes prior-annealed, in order that semi-machining may be executed, either with a view to reducing mass effect for more efficient quenching, later reducing cost through less weight of metal to be heat-treated, or reducing time (and naturally cost) by part-machining in a soft condition.

In the absence of air-hardening propensities, such as in the straight-carbon types, S76 or En 8, where annealing may be required, the Ac2 point at roughly 780°–800° C., depending on manganese, sulphur, etc., is likely to yield the most favourable results. With straight-nickel types, such as B.S.S. 5005/103, 5005/104, S67, En 37, however, normalising alone is usually sufficient, but some machinists prefer annealing, and 730°–750° C. is usually most preferable. It should be clearly understood in all these references that the operation of annealing implies furnace cooling, and not in air, as is the case where normalising is carried out.

Turning now to the question of the die-hards of the forging industry, S82, S90, 2S28, and En 36, we come to the types which are really difficult to obtain in uniformly soft condition. In each case the most suitable temperature is that of recrystallisation at around the Ac1 point, and whilst the actual figure could be determined dilatometrically, it can also be worked out in a few moments from the analysis obtaining. The latter usually yields a figure between 650° and 700° C., and beyond that the question of soaking time and temperature of withdrawal are of importance.

Cases of failure on 2S28 are legion, particularly when a customer asks for a maximum Brinell hardness numeral of 255 when specification allows up to 285. Not all compositions of 2S28 are capable of such low readings, and no length of soak or cooling conditions can avail. The circumstances are made worse if the forgings are only a few pounds in weight, as they tend to cool more quickly in the furnace. One still hears occasionally of the brighter intellects who "take the load a bit higher" in order to make it "come softer," little realising that in a single hour of such they may destroy the work of twenty-four.

Undue distortion during carburising sometimes takes place on machined alloy steel components. One of the two following alternatives is frequently employed to obviate the difficulty. Firstly, the forgings are soaked for a minimum of 2 hours at a higher temperature than that at which carburising is carried out, usually around 1,000° C. The charge is allowed to cool in furnace to the annealing temperature of 650°–700° C. and then annealed in usual manner. Secondly, the forgings may be oil-quenched from 850° C. and then annealed in normal manner. Both methods are claimed to get rid of all "movement" prior to the case-hardening stages after machining.

Quenching (for Hardening)

Those friends who are familiar with heat-treatment will doubtless forgive the qualification contained in the sub-title, as not all efforts of quenching result in the occurrence of hardening.

The most-used range of heat-treated steels is without question 55/65 tons in tensile. Until quite recently seven or eight different types met this requirement, with probably 4S11 as the most popular. The last-named specification consists of a low carbon nickel-chromium-molybdenum type which has fairly good air-hardening properties. Providing the cast material is uniform, there should be little difference in hardness values obtained on quenching in oil as compared with air cooling, figures of 415/477 B.H.N. (3·0/2·80 mm.) being typical prior to tempering, unless sections are exceptionally heavy—e.g., beyond 4 in. R.T. The result was that for the most part production ran smoothly with few difficulties.

Unfortunately, however, the necessity for conservation of alloying elements has arisen, and certain steels in the En series have been put forward as substitutes. Few of the substitutes have any air-hardening propensities, and this factor, coupled with in one instance a carbon tolerance of 0·25–0·40%, is doubtless the cause of much irregularity during treatment. It may well be imagined that in cases where a furnace load of several tons is quenched automatically by means of a charging machine, the centre portion is scarcely likely to receive the same efficiency of quench as the outer, if the steel lacks air-hardening capabilities.

Beyond the foregoing, the manganese-molybdenum types, En 16 and 17, have been found to lack in the normally expected co-ordination between Brinell readings and tensile strength, and whereas a 4S11 steel of 248 B.H.N. (3·85 mm.) would usually give 55 tons per sq. in. ultimate stress, an En 16T of similar reading might be as low as 48 tons per sq. in., with a proof stress of 37–38 tons instead of 43 minimum.

The phenomenon (?) appears to be one peculiar to the manganese-molybdenum types of steel and has no actual remedy, but the heat-treater should employ a modified range of Brinell readings where the minimum figure is likely to yield the minimum requirements of ultimate and proof stress. Such figure is usually taken as being 269 B.H.N. (3·70 mm.) in the case of En 16T.

The confusion likely to arise over a restricted or modified Brinell range may well be imagined particularly when the actual specification as laid down remains unaltered. The controlling factor, however, is the replaced specification, and the actual physical tests should be observed, using the Brinell readings only as a guide.

Perhaps at this point a word on replacement of specifications may be in order. In the British Standards

Schedule 970 : 1942 for Wrought Steels for General Engineering Purposes (known as the En series) are listed approximately 60 steels designed to replace B.S. 5005, 5008, 5010 and part of 5006, as well as the " S " series of aircraft steels. In addition to this, has been issued an index in which are contained En replacement specifications for the whole of the British Services, railway companies, industrial as well as American schedules. Not only are users expected to specify En series steels, therefore, for the sake of uniformity, but to employ the advised or allowed substitute for the purpose of alloying element conservation.

The correct way of setting down such substitutes, or possibly alternatives, would be a more diplomatic description, is to state first the replacing, or used, material, followed by that replaced contained in brackets—e.g., En 16T (4S11). The interpretation of such would then be that En 16 steel had been used, range " T " of the schedule was required, but 4S11 having been the replaced type, the physical tests of the latter had to be attained.

In explanation of the suffix " T," it should be said that the series of letters N to Z inclusive have been adopted as representing the same lower limit of the tensile range with all specifications. Commencing with symbol N at 30 tons per sq. in., each succeeding letter indicates an additional 5 tons until symbol Y at 80 tons is reached. Beyond this point there is an increase of 20 tons for symbol Z to represent 100 tons minimum.

The lower alloy content of certain of the alternative steels may often mean a higher quenching temperature than that necessary for the replaced specification if the same quenching medium be employed, and in some cases it is necessary to resort to more drastic measures by quenching in water. In the latter instance, a lower temperature may be advisable to avoid hardening cracks.

Knowledge of extreme cases are available where, even after drastic quenching in water it has not been possible to reach the minimum ultimate stress requirements with certain forgings, due either to a heavy ruling thickness or a particularly low carbon, though the latter may still be within the range of composition allowed.

Determination of correct quenching temperatures may well present another problem. Whilst dilatometric recordings offer the correct technique for the procedure, one can scarcely call a halt on commercial heat-treatment meanwhile a laboratory staff take forgings, the property of other people, and carry out investigations. A ready, but not too rough an alternative consists briefly of an assimilation of the hardening value of the chemical composition. Admittedly, it is necessary to use a cast analysis, but after all, this is just as likely to represent the average of the batch, as a further odd analysis taken from any one given forging.

The iron-iron-carbide diagram caters solely for the alloy of carbon and iron and takes no account of any other elements which may be dissolved in it. As such latter, when present, play quite an important part in their effect on transformation points, adjustments automatically become necessary. Taking first the common elements, the author deducts from the manganese percentage ten times the sulphur present and of the balance of the former element one-third is taken and added to the carbon. Reference to the iron-iron-carbide diagram will now show the desired Ac_3 temperature for the revised " carbon " content. As an example, with a steel containing 0·35% carbon, 0·60% Mn, and 0·045% S, 0·05% of manganese would need adding to the 0·35% C and so making of it a nominal 0·40% C.

Explanation should be made that quite contrary to theory where the molecular proportion of manganese to sulphur in the MnS compound is 1·72 to 1, actually ten times the amount of manganese is always taken in practice as being necessary.

Beyond the three last-named elements, the amounts of phosphorus normally present may be disregarded for heat-treatment purposes as may also silicon, as although the latter does raise the Ac3 point if present in sufficient quantity, the 0·30% maximum of the specification again renders negligible the effect of the small amounts present.

Turning now to the alloying elements—nickel, chromium and molybdenum—let us take a typical 4S11 analysis as shown in Table I.

TABLE I.

TYPICAL AVERAGE ANALYSIS OF STEEL TO SPECIFICATION 4S11.

	%
Carbon	0·31
Silicon	0·19
Manganese	0·54
Sulphur	0·026
Phosphorus	0·025
Nickel	3·19
Chromium	0·71
Molybdenum	0·22

The common elements, dealt with as already described, yield a " carbon " figure of 0·40% for which the Ac3 point would be roughly 820° C. Taking the nickel as depressing the point at the rate of 10° C. per 1% present, we now arrive at 788° C. The effect of chromium and molybdenum is roughly double that of nickel in the reverse direction, which supplies an elevation of 19° C. and so a final figure of 807° C. is attained. But this is the actual point, and due allowance must be made for transference from furnace to quenchant, and taking a nominal 20° C. for such, an ultimate figure of 827° C. is presented. Set forth in above manner, the calculations may sound somewhat formidable ; actually, however, those accustomed to the work run off the whole thing mentally and arrive at the answer in a few seconds.

Alloy steels containing nickel, chromium or molybdenum should rarely if ever be quenched other than in oil, for obvious reasons, but many of the manganese-molybdenum substitutes require water, particularly if the carbon is below 0·30%.

Chrome-molybdenum steels of the En 19, En 20 types appear to make the structure much more sluggish than other specifications in transformation response, and experience indicates that a soaking time of at least an additional 50% is necessary to allow for the complete entry into solid solution of the pearlite eutectoid. Failure to observe this point results in a very coarse structure and low Izod values after tempering.

High N-C-M steels of En 30 (2S28) character are capable of air-hardening from 830° C., and so the normalising and hardening temperature and treatment act similarly in producing the full ultimate stress value of approximately 100/110 tons tensile.

Tempering

No alloy steel on any account should be used without tempering after being quenched for hardening. Specification details are couched in usual vagueness ; frankly,

this must be so because of the wide range of analyses possible and also the character of forging which may weigh from an ounce to a quarter of a ton.

For the higher carbon N-C-M and Ni steels of 4S11 and S69 type, tempering is usually necessary between 600° and 650° C. A wise precaution, however, is to Brinell the material prior to the operation, which may serve as a guide to tempering.

Lower carbon, but higher chromium types, with Ni and Mo present, such as S65, are more frequently dealt with between 550° and 600° C., but the manganese-molybdenums may easily vary from 450° to 680° C. in their requirements.

The foregoing deals with routine working smoothly, and now for the troubles which may be encountered.

Assume that an N-C-M steel of 4S11 type has been tempered at 640° C. and instead of conforming to desired physical requirements, shows hardness value of 341–363 (3·30–3·20 mm. imp.). The natural tendency is to retemper at a higher temperature. This is, therefore, done, but instead of a slightly milder reading, one is surprised to find the forgings *harder !* The reason for this is that the average composition is capable of promoting air-hardening, and whether the forgings are air-cooled or quenched, on withdrawal they are more likely to become harder than softer.

Remedy in a case of this character is to reharden, employing a lower quenching temperature in order to restrict the amount of carbide entering solution on quenching. Forgings in specification 4S11 weighing only one or two pounds may frequently be hardened from as low as 780° C. instead of the more usual 830° C.

The foregoing remedy may also be applied if forgings are liable to crack. Those which are irregular in shape with plenty of lugs, bosses and/or collars, will usually withstand practically any quenching temperature, but one which has a smooth exterior, like a bottle, is prone to split longitudinally. The reduction in temperature as mentioned, however, will invariably cure this.

In the realms of the hierarchy it is firmly believed that a correctly hardened batch or consignment should not only be delightfully uniform in hardness, as measured by Brinell, but suitably centred in the permitted range of the specification. Unfortunately, such high spots are not always attainable, and in such cases it is necessary to carry out 100% Brinell hardness test. Those too soft should be rehardened, probably from a higher temperature and subsequently retempered along with those found too hard.

Mixed steels—perish the thought !—may also be sorted out by Brinelling, unless, of course, they are so similar in composition that the response on heat-treating bears close resemblance.

Quenching after tempering is a usual practice to ensure improved Izod values. Unfortunately, this sometimes results in the formation of a superficial skin of slightly greater hardness which gives a false Brinell hardness reading and may cause to enter service forgings which are below specification requirements after the outer surface has been removed on machining. Care must be taken to avoid such occurrence, omitting, if necessary, the quench.

Miscellaneous Difficulties

Test-pieces differing in size from the batch covered present another difficulty. The difference usually lies in the T.P. being forged smaller than the forging. In consequence of this it is liable to become much harder than the work, which makes it almost impossible for both to come right at the same tempering temperature.

Alleged hardness discrepancies on forgings sometimes occur between the heat-treater and customer. Upon examination this may be found due to a difference in type of hardness testing machine employed. It must be appreciated that the same scale of readings cannot be expected on any given size forging from, say, a 3,000 kg. load and 10 mm. ball as from a 30-kg. load and 1 mm. ball. The indentation of the latter is naturally confined more to the surface and invariably yields a harder reading, particularly if quenching after tempering has been done. Wherever possible, the standard loading of 3,000 kg.–10 mm. should be employed. Beyond this, a 5 mm. ball under 750 kg. is the next practical step. But having decided on the right type machine, all checks should be made subsequently by that means.

Temperature measurement, or lack of it, can cause endless trouble. It should be realised that the whole success of heat-treatment depends entirely on this point. It is necessary, therefore, that the most accurate and reliable of instruments should be employed, and beyond merely measuring the temperature it should be recorded in addition, as well as control full supply through the valve or switchgear.

Low Izod values is a difficulty occurring more frequently with the greater use of substitute steels, particularly those of Mn-Mo type. Values of 20–30 ft.-lb. may frequently be raised to 40–60 ft.-lb. merely by repeating the tempering operation for the same time cycle as before, and water quenching. Doubling the time on the first tempering does not have the same effect.

It may be, of course, that for some undue reason the normal treatment cycle has not been able to obtain the expected quenching response, in which case a special prior-normalising is sometimes effective.

The Nickel Bulletin, No. 10

THE "Nickel Bulletin" for October includes abstracts on complex nickel-aluminium-iron magnet alloys, the effect of hydrogen on ferrous castings, weldability tests, the use and misuse of the salt-spray test and base metal dental alloys for war-time use in place of gold alloys.

Copies of the "Nickel Bulletin" may be obtained, free of charge, from the Bureau of Information on Nickel, the Mond Nickel Company, Ltd., Grosvenor House, Park Lane, London, W. 1. Individuals are requested to state the name of their firm or otherwise indicate the nature of their interest.

We are informed by High Duty Alloys, Ltd., of Slough, that the activities and control of their subsidiary company, Hiduminium Engineering, Ltd., have been changed, and the parent company has formed a new department to continue the work formerly carried out by Hiduminium Engineering, Ltd. Mr. J. A. Heron, formerly managing director of Hiduminium Engineering, has severed his connection with the company.

The Late Dr. W. H. HATFIELD, F.R.S.

THE death of Dr. W. H. Hatfield, briefly announced in our last issue, at his home in Sheffield, on October 17, after a very short illness, was a great blow, not only to his wife and relatives, but to his many friends and to the iron and steel industry as a whole. Born in Sheffield on April 12, 1882, it can be said that he devoted his life to the advancement of Sheffield's primary industry.

Educated at Sheffield University College and at the Applied Science Department of Sheffield University, he gained the Mappin Medal in 1902. His first published researches, on "The Elimination of Silicon in the Acid Open Hearth," were carried out in collaboration with the late Professor A. McWilliam, at that time Associate Professor of Metallurgy to Professor J. O. Arnold, and presented before the Iron and Steel Institute in 1902. A second paper, on "Acid Open Hearth Manipulation," again in collaboration with Professor McWilliam, was presented to the Institute in the autumn of 1904.

As a result of these researches, he was awarded the Carnegie Research Scholarship by the Iron and Steel Institute in 1905, and this he devoted to a study of the "Condition of the Several Varieties of Carbon upon the Strength of Cast Iron as Cast and Heat Treated," the results being presented to the Iron and Steel Institute in 1906. It is noteworthy that other recipients of the Scholarship that year included L. Guillet (Paris), W. Rosenhain and H. C. H. Carpenter. The award of a second Carnegie Scholarship, in 1906, enabled him to continue his study of cast iron, the results of which were presented in his fourth paper to the Iron and Steel Institute in 1907.

At this period of his career he was associated with the firm of Messrs. J. Crowley and Co., Ltd., and in 1908, in a paper presented to the Institution of Engineers and Shipbuilders in Scotland, he described the production of malleable cast iron giving 19% elongation, associated with a maximum stress of 23 tons per sq. in. The achievement represented by these values will be appreciated when it is borne in mind that the normal elongation up to that time had been of the order of 3 to 6%.

In 1913 the degree of Doctor of Metallurgy was conferred upon him, and three years later he joined the firm of Thos. Firth and Sons, Ltd., where he took over the direction of the Brown Firth Laboratories in succession to Mr. Harry Brearley, the inventor of stainless steel, and later was appointed a director of Thos. Firth and John Brown, Ltd., and of Firth Vickers Stainless Steels, Ltd., positions he held at the time of his death.

From 1913 until the end of 1918 he devoted much of his attention to the application of the stainless and other alloy steels to aero-work, although in 1917, at the request of the Foundry Committee of the Iron and Steel Institute, he prepared a report on the position of the malleable castings industry in this country for the consideration of the Ministry of Munitions. At the end of hostilities of the last Great War, he turned his attention to the development of the stainless steels, and in 1924 first placed on the market the austenitic chromium-nickel steels of the 18–8 type. Later, he found that steels within a certain narrow range of composition—namely, 11·5 to 13% chromium and 11·5 to 13% nickel, possessed excellent resistance to corrosion with increased malleability, rendering them particularly suited for deep-drawing applications. Next followed an intensive research to overcome the difficulties presented by the susceptibility of the 18–8 type of steel to intergranular corrosion, when subjected to chemical attack after welding or hot-working. His researches on "creep" resistance resulted in the development of several types of heat-resistant steels which to-day are such an important factor in the construction of high-power engines.

In October, 1928, he presented the Campbell Memorial Lecture before the American Society for Steel Treating, in Philadelphia. In 1933 he was awarded the Bessemer Gold Medal, and in making the presentation, the then President of the Iron and Steel Institute, the late Mr. W. R. Lysaght, C.B.E., concluded his appreciation of Dr. Hatfield's work up to that time in the following words: "In a great measure, Dr. Hatfield had succeeded in co-ordinating and developing co-operative research within the industry, and had promoted a general spirit of collaboration which was so important a factor in industrial progress. In pursuing his scientific work he had ever kept the practical end in view, and he had fully justified his calling as a scientific metallurgist by the success which he had obtained by the introduction of new commercial products on a large scale."

There can be no doubt that Dr. Hatfield's greatest contribution to the iron and steel industry has been his work in the inauguration and co-ordination of the

research organisations of the country. He stimulated the formation of the Joint Research Committees of the Iron and Steel Institute and the British Iron and Steel Federation, reporting to the Iron and Steel Industrial Research Committee. In 1924, he was asked to form the first of these Committees—the Heterogeneity Committee—and he has been its chairman since that time. He was also chairman of the Corrosion Committee and of the Alloy Steels Committee and of many sub-committees.

In view of his wide experience and activities, it was natural that he should be appointed chairman of the Technical Advisory Committee, constituted in May, 1940, by the Special and Alloy Steels Committee, with a mandate to deal with the rationalisation of alloy steels and to effect the necessary economies and adjustments in the use of ferro-alloys. In his capacity of chairman of this Committee he was a member of the British Mission which visited the United States of America in the spring of this year, with the object of co-ordinating the resources of the United Nations. It was this subject, to which he had devoted so much effort, which he chose as the theme of the paper he presented to the British Association in September, 1942.

He was a member of the Iron and Steel Industrial Research Council and chairman of its Programme and Finance Committees. He served on the Executive Committee of the National Physical Laboratory, and was also a member of the Metallurgy Research Board of the Department of Scientific and Industrial Research.

Of his characteristics, his dynamic energy was the most outstanding. Alert and brisk in all his actions, he appeared to have unlimited vitality and got through an incredible amount of work. He never spared himself, and seldom took a holiday, even in the days of peace, but nevertheless remained astonishingly young. This was probably due to his mental attitude, for he regarded life as a game which he played hard and strenuously, but with considerable zest. He had no use for glibly accepted theories, and loved debate in which he was often deliberately provocative, but a twinkle in his eye would show that his object had been achieved when his opponent, goaded into action, had to acquire evidence to prove his case.

He was the most generous of men in all respects; there was nothing petty about him. He was always ready to give an opportunity to the young men to show initiative and in return he inspired not only enthusiasm but affection. Casual acquaintances may have been misled by a mask of cynicism he was apt to assume, but those who had been privileged to have his friendship, or had had cause to experience his ready sympathy, were not deceived in this way. We can speak sincerely of his generosity, because, since the inception of this journal, we have frequently sought his advice and help on editorial matters, and, although he was an exceedingly busy man, his counsel was never sought in vain; he always responded in a most generous and helpful manner.

No one had been more deeply shocked than he over the sufferings of the people during the air raids of 1940-41, and it had been his purpose to devote the remainder of his life to the reconstruction of his native city. It was typical of him that when, on a certain morning in December, 1940, he found the Brown Firth Research Laboratories blasted by enemy action, he entered the unsafe building without hesitation, inspected the damage, and on his own responsibility got the builders at work almost at once.

He was elected a Fellow of the Royal Society in 1935. In addition to the Iron and Steel Institute, of which he was a Vice-President, he was a member of the Royal Institution of Great Britain, Fellow of the Royal Aeronautical Society, member of the Institution of Mechanical Engineers, the Institute of Physics, the Institute of Metals, the Faraday Society, the American Society for Metals and the American Society for Testing Materials, and had received the Compton Medal from the Institution of Automobile Engineers. He was also founder and first President of the Sheffield Metallurgical Association, and at the time of his death was President of the Sheffield Society of Engineers and Metallurgists.

His publications include "Cast Iron in the Light of Recent Research," "The Application of Science in the Steel Industry," and more than one hundred and fifty scientific papers presented to various Societies, including the Royal Society, the Royal Institution, the Iron and Steel Institute, the Institution of Mechanical Engineers, the Institute of Metals, the Royal Aeronautical Society, the Institute of Automobile Engineers and the American Society for Metals.

Britain and, in particular, Sheffield has lost a son of outstanding ability, and they, as well as his many friends in widely varying spheres of life, in all parts of the world, are the poorer by his death.

J. J. A. Jones, M.Sc., F.Inst.P.

We regret to record the death on November 1 of John Jenkin Ashton Jones, of the Central Research Department of the United Steel Companies, Ltd., Sheffield, and well known in metallurgical research circles. Born in May, 1894, he graduated from Aberystwyth in 1915. Shortly afterwards he joined the Royal Welch Fusiliers and saw active service in the Great War, during which he was awarded the Military Medal.

In 1917 his military service was terminated as the result of wounds, and he joined the Research Department at Woolwich, where he assisted Dr. Greaves. In conjunction with the latter, he presented four papers before the Iron and Steel Institute, of which he became a member in 1920. He was co-author of one other paper, and in his own name contributed six further papers.

In 1930 he joined the United Steel Companies, Ltd., and, with Dr. Swinden, Director of Research, he took a large part in planning the Central Research Department of that group at Stocksbridge. When the Central Research Department opened he became the manager, a post he held to his death.

John Jones represented the best type of scientific worker, and the quality of his scientific papers needs no elaboration here. He did much other valuable work which was not so well known. His quiet yet forceful personality and the multiplicity of his contacts gained him many valuable friends, to whom his many other qualities endeared him. As an example of his modest self-effacing personality, it may be noted that many who knew him intimately were quite unaware of his distinguished military record.

His death will be deeply felt by all who were in contact with him in personal, business and scientific relationships. We join with his many friends in expressing our deep sympathy to Mrs. Jones and family.

The Use of Basic-Lined Ladles in the Desulphurisation of Cast Iron by Sodium Carbonate

Sodium carbonate has been used for many years to desulphurise cast iron in siliceous lined ladles; during recent years, however, there has been a demand for a greater degree of desulphurisation than appears to be possible with soda ash in siliceous lined ladles, and, in a recent paper, N. L. Evans describes experiments on the use of basic linings with the sodium carbonate process, from which some results are abstracted and given here.*

THE use of sodium carbonate for desulphurising cast iron has increased steadily over the past twelve years or so, until at the present time the quantity of metal treated probably exceeds two million tons per annum. The process is normally carried out without the use of any special materials for lining the ladles in which the reaction takes place. Ladles for molten iron are commonly lined with acid (siliceous) refractories, which may consist of sand, ganister or similar substance, rammed around a former or daubed on as a stiff paste to form a monolithic lining, or else of brickwork, usually firebrick containing 35–40% of alumina and the balance mainly silica, the joints between the bricks being a cement similar in composition to that of the bricks.

It has long been known that the reaction whereby sulphur is removed from iron by sodium carbonate may be reversed in the presence of an excess of silica in the desulphurising slag according to the equation:—

$$Na_2S + SiO_2 + MnO = Na_2SiO_3 + MnS$$
$$(FeO) \qquad\qquad (FeS)$$

The use of sand for thickening the slag to facilitate its removal is undesirable therefore, and in its place limestone, either in the form of a powder or as dust-free chippings, is used.

Soda slag which has been used for desulphurisation usually contains upwards of 30% of SiO_2, this being derived from the ladle refractory, from furnace slag, which is often allowed to run into the ladle in spite of efforts to prevent it, and from silicon which is to some extent removed from the metal by the action of the soda slag. This degree of contamination of the slag does not matter in the case of most of the iron which is treated by sodium carbonate, and which contains before treatment 0·06–0·15% of sulphur. By the use of 1% of sodium carbonate about half this sulphur is removed, giving the metal a sufficiently low sulphur content for most industrial purposes. It is a more difficult problem, however, to effect an appreciable reduction in the sulphur content if the iron contains initially not more than 0·04% of sulphur and less than 1% of silicon. The effect of treating such an iron with even 1½% or 2% of sodium carbonate may be to reduce the sulphur by not more than 20–25%. Irons containing relatively low percentages of manganese are also difficult to desulphurise.

During the past few years a demand has arisen for a greater degree of desulphurisation of cast iron than is possible by the application of sodium carbonate in ladles with siliceous linings. The following are some of the problems which have been encountered:—

(1) In the manufacture of recarburising iron the raw material is the highest grade of hæmatite pig iron. This usually contains not more than 0·025% of sulphur, and it is desired to reduce this to 0·010% or less.

(2) In the production of hæmatite pig iron it is desirable to reduce the sulphur to the lowest possible limit, provided that the phosphorus content is also low.

(3) Owing to the war-time shortage of pig iron and, pre-war, the abundance and cheapness of steel scrap, there is a strong inducement to use up to 95% of steel scrap in the cupola charges to produce metal for electric steel furnaces, for Tropenas steel castings and for refined pig iron.

(4) In order to conserve certain alloying elements such as nickel, chromium and molybdenum, the "Rowden" process has been revived during war-time. This process involves the production, in a basic open-hearth furnace, of a synthetic pig iron from alloy-steel scrap, and it is desired to desulphurise this in the ladle after tapping it from the furnace to bring the sulphur below 0·02%.

(5) In the production of low-silicon basic iron for steel-making it is now becoming necessary to use iron ores containing more sulphur than was usual in the past. Owing to the better-grade ores having been used up, this difficulty is likely to be accentuated in the future. It is often necessary to remove more than 50% of the sulphur from the iron by a single sodium-carbonate treatment, particularly when the "acid burden" process is used.

One possible method of improving the efficiency of the sodium-carbonate process to solve the above problems is to substitute basic for acid materials in the refractory linings of the ladles used, so eliminating one source of silica in the desulphurising slag.

This is not an entirely novel suggestion, though the author is not aware that it has previously been extensively practised in Great Britain. In Luxemburg,[1] before the outbreak of war, ladles lined with tar/dolomite were used for the desulphurisation of basic Bessemer iron with sodium carbonate, and it is reported that the lives of the linings ranged from 800 to 1,000 heats, as compared with 400 to 600 heats in the case of

* Iron and Steel Institute, October, 1943 (Advance copy).

1 N. L. Evans. *Foundry Trade Journal*, 1941, **64**, pp. 19–22.

firebrick linings. When not in use, these linings were kept hot over a blast-furnace gas burner. At a German works,[1] where sodium carbonate for desulphurisation was pre-melted, a rotary furnace lined with tar-bonded dolomite was used for this purpose. After treatment, the iron and soda slag were separated from one another by pouring them into a syphon ladle, this also being lined with tar-bonded dolomite. At another German works[2] experiments were made with pig-iron ladles lined with rammed dolomite, rammed magnesite, and rammed dolomite with 10% of graphite added. The blast-furnace runner was also lined with dolomite, though this is stated to have caused some trouble on account of cracking. The percentage of sodium carbonate used is not stated, but desulphurisation was increased from 32·7% in a fireclay lining to 48·2% in a basic lining.

The present paper describes experiments which have been carried out during the past three years to find in greater detail how the use of basic linings affects the sodium-carbonate process. The experimental work carried out comprised two parts: Small scale work in the laboratory; and full scale work in foundries and in iron and steel works.

Experimental Work

The laboratory work was carried out in a Metropolitan-Vickers high-frequency induction furnace and was concerned with trying out the action of soda slags on various refractory materials and finding how the use of different refractories affects the degree of desulphurisation of certain types of cast iron.

Following many preliminary experiments and as a result of experience gained, a series of melts was planned under conditions as nearly constant as possible, to compare the action of sodium carbonate on various types of cast iron in the four available refractory linings. The refractories used were calcined dolomite, tar-bonded; stabilised dolomite, water-bonded; serpentine, water-bonded; and loam, water-bonded.

Four melts were made in each lining, each melt being treated with approximately 2% of granular sodium carbonate. The melts were made in the following order:—

Melt 1—Grey hæmatite pig iron.
" 2—Grey hæmatite pig iron.
" 3—White hæmatite pig iron.
" 4—Common grey iron.

Two melts of grey hæmatite iron were made because, in the first melt in a lining, the soda slag is almost entirely absorbed by the lining unless the percentage of sodium carbonate is fairly high. The first melt in the series described was, therefore, primarily to seal the pores of the refractory.

The analysis of the irons as received was:—

	T.C.	Si.	S.	P.	Mn.
Grey hæmatite	3·86	2·57	0·019	0·035	0·87
White hæmatite	3·12	0·14	0·029	0·032	0·18
Common grey	3·43	1·66	0·102	1·49	0·67

The results obtained from this series of tests are given in Table I. They show that:—

(1) In the various refractory materials tested, the efficiency of desulphurisation with sodium carbonate is in the following order: Calcined-dolomite/tar; stabilised dolomite; serpentine; loam (sand).

(2) Sodium carbonate gives a worth-while degree of desulphurisation in both calcined-dolomite/tar and stabilised dolomite, even when the initial sulphur content of the iron is low (0·03% or less). With such irons, very little desulphurisation takes place in serpentine or sand linings.

(3) An appreciable degree of desulphurisation takes place in all the types of lining tested when the initial sulphur content of the iron is of the order of 0·1–0·2%.

RESULTS OF DESULPHURISATION WITH 2 % OF SODIUM CARBONATE IN FOUR DIFFERENT REFRACTORY LININGS.

Lining Material:	Calcined Dolomite/Tar.		Stabilised Dolomite.	Ser-pentine.	Loam (Sand).
Grey Hæmatite Pig Iron—					
Sulphur, %: Before	0·019	0·019	0·019	0·019	0·019
After	0·008	0·012	0·013	0·020	0·021
Percentage of sulphur removed	58	36·8	31·5	Nil	Nil
Silicon, %: Before	2·57	2·57	2·57	2·57	2·57
After	1·96	1·95	1·98	2·11	2·30
Percentage of silicon removed	23·7	24·1	23	17·9	10·5
Grey Hæmatite Pig Iron—					
Sulphur, %: Before	0·019		0·019	0·019	0·019
After	0·008		0·012	0·015	0·018
Percentage of sulphur removed	58		36·8	21	5·3
Silicon, %: Before	2·57		2·57	2·57	2·57
After	2·11		2·04	2·26	2·30
Percentage of silicon removed	17·9		20·6	12	10·5
White Hæmatite Pig Iron—					
Sulphur, %: Before	—		0·029	0·029	0·029
After	—		0·009	0·014	0·024
Percentage of sulphur removed	—		69	51·8	17·2
Silicon, %: Before	—		0·14	0·14	0·14
After	—		0·06	0·06	0·10
Percentage of silicon removed	—		57	57	28·5
Common Grey Iron—					
Sulphur, %: Before	0·152		0·152	0·152	0·152
After	0·032		0·059	0·083	0·078
Percentage of sulphur removed	79		61	45·4	48·6
Silicon, %: Before	1·75		1·75	1·75	1·75
After	1·14		1·10	1·15	1·21
Percentage of silicon removed	34·8		37·1	34·2	30·8

After the series of experiments, the dolomite/tar and stabilised-dolomite linings showed practically no sign of attack by the soda slags. Both the serpentine and loam linings, however, were severely attacked and were cut by very deep slag lines.

Several full-scale experiments have been carried out by firms using the sodium-carbonate process for desulphurising molten iron, to which the author makes brief reference. Much useful information has, however, been accumulated, demonstrating the advantage of basic-lined ladles for this process, due to the absence of chemical attack on the lining by soda slags and the improved degree of desulphurisation. The author's conclusions resulting from his experiments indicates scope for further investigations on the subject.

Conclusions

As a result of these experiments the author concludes that the use of basic-lined ladles for carrying out the sodium-carbonate process for desulphurising cast iron results in an appreciably greater degree of desulphurisation than is obtained under otherwise similar conditions in siliceous refractory linings. The improvement is most marked in the treatment of cast iron having initially a sulphur content less than 0·04%. Such iron is sometimes difficult to desulphurise to an appreciable extent in siliceous-lined ladles, but in basic linings it has, in some cases, been possible to reduce the sulphur content to a figure below 0·01%.

Suitable basic lining materials are magnesite, dolomite and stabilised dolomite. The first of these being in short supply, attention was directed mainly to the other two. The best results were obtained in linings of tar-bonded dolomite. Stabilised dolomite, containing about 15% of silica, is not quite so good, but it gives results superior to those obtainable in firebrick, loam or other acid

2 N. Theisen, *Stahl und Eisen*, 1938, **58**, pp. 773–778.

linings, and it has the advantage over unstabilised dolomite that it can be kept in stock without deterioration, by users who have no proper facilities for mixing tar and dolomite and require only small quantities of lining material at a time.

The technique of preparing monolithic basic linings appears to be fairly straightforward. There is scope for further investigation on the subject of basic-brick linings. The question of spalling appears to be less serious in the presence of sodium carbonate slag than the results of the standard spalling test would suggest. Methods of patching linings after a period of use are not so successful as could be desired, and it is felt that this matter would repay further investigation.

Correspondence

The Training of Female Analysts for Steelworks Laboratories

The Editor, METALLURGIA.

Dear Sir,

We have read with great interest the article by Messrs. Thompson and Belcher which appeared under the above heading in your July issue.

A similar scheme of training has been in operation at the County Technical College, Wednesbury, for a considerably longer period than that of your contributors. This course, also under the auspices of the Ministry of Labour and Board of Education, was inaugurated in August, 1941, since which time over 100 girls have been trained and placed.

Examination of the data supplied by your contributors reveals points on which our respective schemes differ and some information and observations may therefore be of interest. The following table summarises the essential differences between the Rotherham and Wednesbury schemes of work.

Item.	Rotherham.	Wednesbury.
Nature of intake	Intermittent in groups..	Continuous stream—entry at any time.
Duration of course.....	12 weeks of 33 hours = 396 hours.	10 weeks of 39 hours = 390 hours.
Time devoted to work other than steel analysis.	17 hours per week throughout the course = 204 hours.	8–11 days of 7 hours at commencement = 56–77 hours.
Time spent on ferrous analysis.	192 hours	313–334 hours.

It will be noted, at the outset, that the courses exhibit a striking disparity in the proportions of time allotted to ferrous analysis and general chemistry. Messrs. Thompson and Belcher have explained in their article the reasons underlying their generous allocation of time to general chemical principles, and it may be of interest therefore to make known why a different procedure has been adopted at Wednesbury.

The urgent need in our local area was, and for that matter still is, for trained girls to replace the junior operators normally employed on the routine analysis of irons and steels. This need appeared to demand a strictly utilitarian policy and the fundamental decision taken at the beginning of the course, and still adhered to, was that the primary object should be to confer upon the trainee a measure of skill in routine ferrous analysis that would enable her to be of immediate value to industry. Academic experience naturally inclined us to a syllabus which would have included a generous measure of theoretical and practical chemistry; on the other hand, from our industrial experience, we were well aware that the bulk of the routine work in industry is performed satisfactorily by boys who are very largely ignorant of the theory underlying the methods which they unthinkingly, but none the less successfully, employ. In their case a knowledge of general chemistry has to be acquired subsequently through the medium of normal evening classes at a Technical Institution. Consequently, we decided to omit all but the barest minimum of general chemistry and to concentrate almost wholly on practical ferrous analysis.

As to the wisdom of this decision, as far as the needs of our local area are concerned, we have no doubts. The welcome given to our trainees in the local works' laboratories, coupled with the fact that firms after taking one girl are anxious to engage others, has amply satisfied us upon this point.

We were glad to note that Messrs. Thompson and Belcher share our views that it is quite impossible to teach other than the barest outline of elementary chemistry in the time available. This makes it difficult for us to appreciate the syllabus of exercises in practical chemistry upon which they have embarked. It seems paradoxical, to say the least, that—having admitted the futility of attempting to impart other than a modicum of chemical theory—they should then see fit to draw up a syllabus of practical work considerably more comprehensive than that usually taken by an Inter B.Sc. candidate. Quite frankly, we doubt the profit accruing to the student who has to scamper through such a syllabus in the short space of 132 hours. We would, however, agree with Dr. Swinden's remark in your August issue, that "such training is amply repaid in the case of those pupils who are likely to succeed." Since these will, presumably, be confined to an exceptionally able minority, it would be interesting to learn how the larger proportion of more mediocre intellects react to work which they must feel to be, both in nature and quality, beyond their capacity to master.

We feel strongly—in the light of our greater experience—that, if we at Wednesbury had so extended the scope of our training, a substantial number of our trainees would have fallen by the wayside and would not now have been performing profitable work in industry. If our Rotherham colleagues take an opposite view we can only be presumptuous enough to assume that they are basing their opinion upon the exceptional, or upon insufficient evidence obtained during the teaching of a restricted number of trainees.—Yours truly,

H. A. MACCOLE.
W. V. SHAW.

County Technical College,
Wednesbury.
October 28, 1943.

The Editor, METALLURGIA.

Dear Sir,

With regard to the assertion that we believed the course of training at Rotherham to be the only one of its kind at the time of its commencement, this information was given to us by the officials of the Ministry of Labour concerned directly with its inauguration. Although your correspondents take some pains to show

that the course at Wednesbury is essentially different, nevertheless if priority is required we have no wish to dispute their claims, but on the contrary congratulate them on tackling the problem so early and with, evidently, so much success.

We do not feel that they have any need to justify at such length their scheme of training. Whatever their scheme is (and unfortunately they give no information on this point) we feel sure their results have justified it. With regard to the syllabus in practical chemistry carried out at Rotherham, we fail to appreciate your correspondents' evident dislike of it. It was stated—one would have thought, reasonably clearly—in the original article why this was devised—namely, to give practice in manipulation of various chemical operations, of volumetric and gravimetric determinations, use of indicators etc. It was not to teach the chemistry of the processes involved. Why the more mediocre intellects should be any more capable of mastering, say, the determination of copper in steel than in copper sulphate, we find it difficult to follow. There is no question of the chemical analyses being "in nature and quality" more intricate and difficult than the steel analyses; in fact, an Honours graduate in chemistry would probably not have any experience in steel analysis.

There is evidence that the course at Rotherham has stimulated the interest of the trainees. Of the 41 now trained, 12 were placed with a factory in a locality not conveniently situated for the continuance of study. Of the remainder, 15 are attending evening classes; 7 are taking the Hons. Chemistry or A.I.C. course, and 8 the National Certificate or City and Guilds' courses. Miss A. S. Lockheart, the Secretary of the Women's Engineering Society, will probably be interested in this information.

With regard to your correspondent's last paragraph, that a substantial number of trainees at Wednesbury would have fallen by the wayside had the scope of their training been extended, they are, of course, better qualified to express an opinion on that than we are.—Yours, etc.

M. B. THOMPSON.
R. BELCHER.

Department of Metallurgy and Chemistry,
College of Technology and Art, Rotherham.
November 8, 1943.

Copper Development Association

Tenth Annual General Meeting

THE tenth annual general meeting of the Copper Development Association was held in London on October 27, and was attended by representatives of member companies. Owing to the unavoidable absence of the chairman of the Council, Mr. D. Owen Evans, M.P., the chair was taken by Mr. E. Pam, O.B.E.

In his speech, the chairman pointed out that this year's meeting marked the completion of the Association's first decade of successful operation, and that it was interesting to recall that when the Association was formed in 1933 it was one of the first of its type in this country. It had proved its value and it was significant that since its inception a number of other Associations on substantially similar lines had come into being, testifying to the increasingly widespread realisation of the need for co-operative effort in development and the essential soundness of the basis on which the Association had been built up and operated.

During the past ten years the Association had emerged from the experimental stage and could now be regarded as an institution that had come to stay, the need for which would in the future most certainly be greater than ever before. Copper would undoubtedly be faced with greater competition from other materials than in any period hitherto and would require the backing of all the research and development that could be brought to bear. In addition, many set-backs resulting from war-time conditions would have to be made good.

Since the outbreak of war the Association's activities had of necessity been greatly curtailed, but an active nucleus had been maintained which was capable of quick expansion. It was necessary that vigorous expansion should take place at the earliest opportunity and that the activities of the Association should be developed to an extent even greater than that prevailing before the war.

He hoped that members and others would continue to take an active interest in the Association and to contribute in every way possible to the furtherance of its objects, so that it might continue to prosper and increase in strength, prestige and usefulness to the benefit of the whole industry.

A Select Bibliography of Foundry Practice

THE Sheffield City Libraries continue to produce, at varying intervals, their Research Bulletins, which are proving so useful. The more recent one is a bibliography on foundry practice, and is an extensive revision of a list issued last year; it has been compiled for the Foundry Society of the Sheffield Trades Technical Societies. This bibliography comprises English and American publications only, and in order to limit its size no articles from technical journals have been listed, but details of English and American periodicals dealing wholly or in part with foundry practice are included. Attention is especially directed to abstracts of technical literature, such as those published by the Iron and Steel Institute, the Institute of Metals, the British Cast Iron Research Association, and the American Foundrymen's Association.

Some of the older textbooks have been included in the bibliography. It is explained that these textbooks frequently provide the only printed record of foundry theories and practices which have been only partially superseded; they are also useful for instructional purposes in classes on foundry practice. This bibliography will be especially useful to readers studying the subject, but owing to the paper shortage, we are advised that copies can only be supplied to firms, technical institutions and research organisations making official application and enclosing threepence to cover the cost of postage. Application should be made to Administration Department, Central Library, Sheffield, 1.

High-Silicon Acid-Resisting Cast Iron*

By J. E. Hurst, D.Met.

The resistance properties of high silicon cast iron against strongly corrosive liquids are discussed and illustrated by diagrams. Attention is directed to its mechanical properties; it is considered that the degree to brittleness of this iron has become greatly exaggerated in the minds of some chemical engineers, due largely to want of care in handling. The manufacture of silicon acid-resisting iron components is briefly considered, and notes are given on the welding of this material.

IT is well established that alloys of iron and silicon containing upwards of 14% silicon possess very great resistance to the corrosive attack of the mineral acids. Because of these excellent corrosion-resisting properties these alloys find a use in the chemical industry where resistance to strongly corrosive liquids, coupled with metallic properties, is essential. These alloys are manufactured and sold under a variety of trade names, but are all embraced in the phrase high-silicon acid-resisting cast iron.

Corrosion Resistance

The corrosion resistance of alloys of iron and silicon in contact with boiling 20% sulphuric acid is illustrated by the diagram, Fig. 1. The rate of corrosion is expressed in millimetres of acid penetration per year, and it will be observed that the magnitude of acid attack decreases rapidly with increasing silicon content. Above a value of 14·25% further increases in silicon do not result in a proportionate increase in resistance. The diagram (Fig. 2) illustrates the resistance of the iron silicon alloys to boiling 70% nitric acid. The effect of this acid is not so violent, and a smaller minimum percentage of silicon can be tolerated. The diagram shows that an alloy containing 12·5% silicon has sufficient resistance to boiling 70% nitric acid. Maximum permissible penetration of 0·1 millimetre per year is accepted for this class of material, and on this basis any alloy containing more than 14·25% silicon can be regarded as having a satisfactory acid resistance. Silicon iron of this composition will withstand successfully all concentrations of these two acids, either hot or cold or in admixture, and on these grounds they find a great many applications in acid manufacturing equipment.

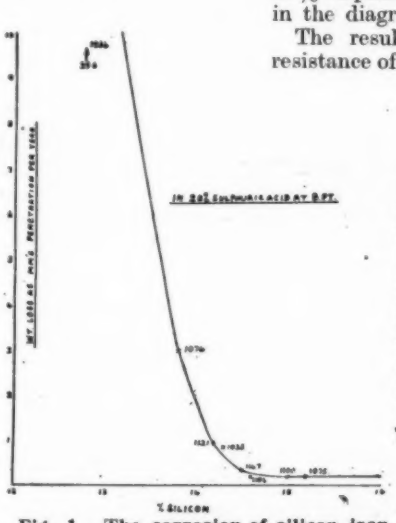


Fig. 1.—The corrosion of silicon iron in boiling 20% H_2SO_4.

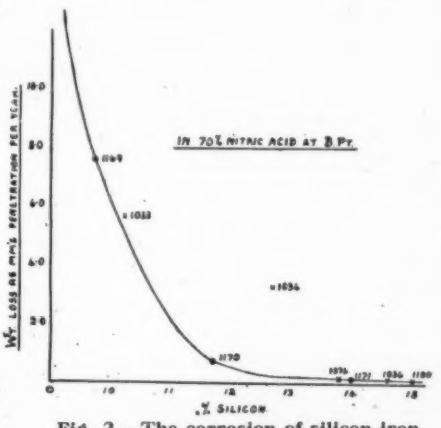


Fig. 2.—The corrosion of silicon iron in boiling 70% HNO_3.

Acid-resisting silicon iron is most easily attacked by the halogens and halogen acids. The diagram (Fig. 3) shows that the attack of hot concentrated hydrochloric acid is appreciable, but in the cold the rate of corrosion is negligible. The resistance of 15% silicon iron to hot hydrochloric acid solutions is increased substantially by the addition of molybdenum. The diagram (Fig. 4) shows the results of an investigation of the corrosion resistance of a 15% silicon iron containing 3·5% molybdenum, compared with a molybdenum free alloy in boiling 70% hydrochloric acid. The corrosion resistance is expressed as loss in milligrammes per sq. centimetre, and for the purposes of further comparison the curve showing the resistance of the 15% silicon alloy to boiling 20% sulphuric acid on the same basis has been included in the diagram.

The results of a determination of the corrosion resistance of silicon iron in phosphoric acid over a period of 100 hours are summarised in the diagram (Fig. 5). The attack on silicon iron by concentrated and dilute phosphoric acids is negligible at room temperatures. At elevated temperatures the attack by both concentrated and dilute acids was of the same order as that obtained in 20% boiling sulphuric acid.

Tests carried out recently indicate that 15% silicon iron has a very good resistance to the hot acid liquors normally used in dyeworks. A polished test bar immersed for 4,000 hours in a dye liquor under factory conditions remained bright and free from any sign of corrosion.

The effect of moist atmosphere conditions on silicon iron is exactly similar to that of strong acids. Below a critical percentage the alloy will rust when exposed to the atmosphere, whereas alloys of the higher silicon percentage remain quite bright and clean. The critical percentage of silicon is rather higher than that for satisfactory resistance to 20% sulphuric acid (boiling),

*Recently presented before the London and Lancashire Branches of the Institute of British Foundrymen.

probably in the region of 14·7% or higher. It is possible to observe acid-resisting silicon iron castings of correct silicon percentage and satisfactory acid-resisting qualities which have rusted on exposure to the atmosphere.

It is a matter of some theoretical metallurgical interest that the critical percentages of silicon in these alloys for corrosion resistances in various media should all lie at concentrations well within the alpha single-phase area—an area of uniform solid solution up to a silicon percentage of 15·7%.

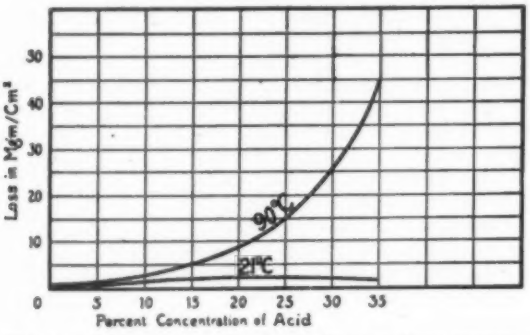


Fig. 3.—Action of hot and cold HCl of various concentrations on Duriron ; 8 hours' test ; agitated solutions.

Mechanical and Physical Properties

The V.D.H. number for commercial varieties is probably lower than that for intrinsic hardness of the alloy. The V.D.H. number of a completely degasified slowly cooled ingot made by melting in a high-frequency furnace *in vacuo* was 520. It is general practice on account of this high hardness to do machining operations by grinding. With suitable machine tools machining can be done with carbide tools. Turning, boring and facing operations can be done readily by either method with suitable equipment. Drilling and tapping operations, however, are not possible, and provision for bolt holes is normally made in the casting.

The mechanical strength of iron silicon alloys is lowered in proportion to the amount of silicon present. Commercial 15% silicon acid-resisting alloy of good quality has a strength value of approximately half that of a good quality grey cast iron. The routine determination of the strength in the author's works is carried out by transverse bending and the moduli of rupture values for 15% silicon iron are of the order of 15 to 20 tons per sq. in. and 20 to 30 tons per sq. in. for the 12% alloy, which is made for certain special purposes.

The degree of brittleness of acid-resisting silicon iron associated with the high hardness and relatively low strength has become wholly exaggerated in the minds of some chemical engineers. Acid-resisting qualities of stoneware, constructional material used extensively in acid-plant construction, has a strength value of only one third that of silicon iron, and, from a hardness and machinability point of view, is subject to similar treatment. Many of the failures encountered in the erection of plant would be avoided by handling silicon iron with care equal to that used when handling stoneware.

The Manufacture of Acid-resisting Silicon Iron

Certain chemical processes in acid manufacture call for equipment which will withstand highly corrosive conditions. In the manufacture and the concentration of nitric and sulphuric acids, dephlegmators, columns, evaporating pots, coolers and condensers are required to withstand these acids in all concentrations and temperatures up to 250° C.

The chemical composition of acid-resisting silicon iron, as made by different manufacturers, varies fairly widely in certain particulars and the melting procedure and technique differs in different foundries. The cupola, crucible reverberatory, open-hearth and electric furnaces are all used. In the factory with which the author is connected electric arc furnaces are used.

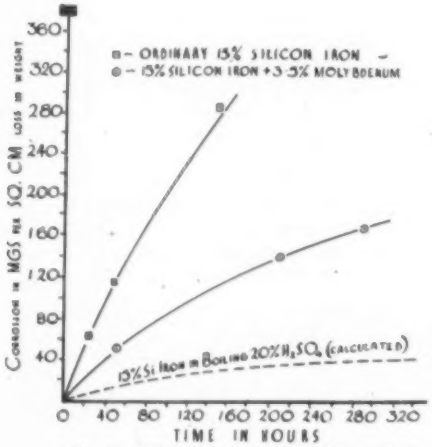


Fig. 4.—The corrosion of silicon iron in boiling 70% HCl.

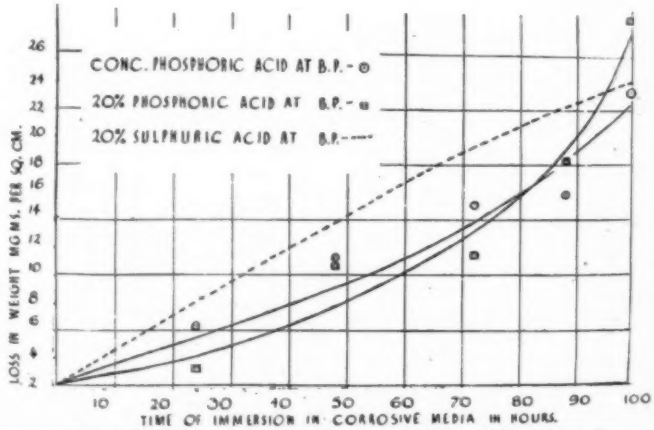

CONC. PHOSPHORIC ACID AT B.P. – ○
20% PHOSPHORIC ACID AT B.P. – □
20% SULPHURIC ACID AT B.P. – ----
LOSS IN WEIGHT MGMS. PER SQ. CM.
TIME OF IMMERSION IN CORROSIVE MEDIA IN HOURS.

Fig. 5.—The corrosion resistance of 15% silicon iron in phosphoric acid over a period of 100 hours.

The Importance of Carbon Content.—The commercial silicon iron alloys contain carbon, and, as in the case of grey iron, the carbon content has an important influence on the founding properties of the alloy. The presence of carbon in commercial silicon iron converts the alloy into the eutectiferous type. From experimental results in the foundry an approximate equilibrium diagram has been

plotted. This is an approximate plane section of the ternary or more complex system, but it serves to indicate results of practical importance. Fig. 6 shows this diagram, which discloses the equilibrium conditions for an iron containing 15% silicon and 0·65% manganese over a range of carbon contents of 0·3 to 0·9%. The appearance of the graphite in silicon iron is apparently some indication of whether the alloy is hypo- or hyper-eutectic. Hypo-eutectic alloys have a " fine graphite "

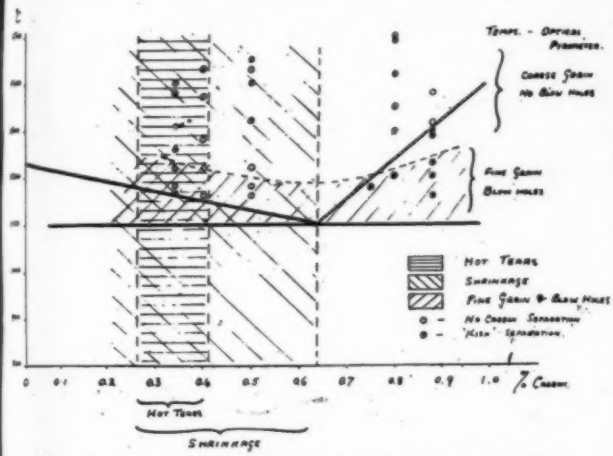


Fig. 6.—Commercial iron-carbon diagram for 15% silicon iron.

type of structure as shown in Figs. 7, 8 and 9. The quantity of " fine graphite " becomes larger as the eutectic carbon percentage is approached. Hyper-eutectic irons contain in addition to " fine graphite " quantities of primary or flake graphite, as shown in Fig. 10. If a molten hyper-eutectic alloy is stirred whilst cooling down, " kish " separates, and the resultant casting is lower in carbon. Figs. 11 and 12 show the micro-structure of the hyper-eutectic alloy before and after stirring or " puddling." The graphite flakes appear slightly smaller in the " puddled " iron, but it is noteworthy that the percentage of carbon is not reduced to quite the eutectic value.

In the alloy of this silicon and manganese content with the carbon content at 0·50% to 0·6% sounder castings with less tendency to hot-tear defects are obtained. The persistence of a carbide phase in the " as-cast " alloy will be mentioned later under the heading of heat-treatment.

The Importance of Gas Content.—The greatest single trouble in the melting of silicon iron is in ensuring that the metal is sufficiently free from dissolved gases to produce sound castings, and precautions need to be taken at every stage of the entire manufacturing operation to guard against the accidental introduction of hydrogen-bearing materials into the molten metal. From vacuum fusion methods it has been established that the presence of CO and H_2 play the most important part in the production of porosity and unsoundness. The total quantity of hydrogen in all the raw materials when in a perfectly dry state was not sufficient to account for all the hydrogen found in the castings in some instances, and this observation directs attention to the possibility of moisture being introduced by a wet charge, a new lining or patch in the furnace or ladle. The ferro-silicon has been shown to have the largest hydrogen content of the dry materials charged. For example, the hydrogen content of a specimen of ferro-silicon was found to be 9 mls. per 100 grms., whilst the hydrogen content of a good silicon iron casting was 2 mls. per 100 grms.

The molten metal looses hydrogen slowly to the furnace atmosphere, and the progress of refining during melting can be observed by taking spoon samples at intervals, the extent of the refining action being judged by the extent of the " sink " in a chill-cast or sand-cast bar.

Silicon iron is usually cast in dry sand and moulds, and although successful castings can be made in green sand, the drying of the moulds offers an additional safeguard against the accidental introduction of moisture or hydrogen into the molten metal.

The Importance of Casting Temperature.—The rate of cooling over the solidification range has a big effect upon the grain size of any metal or alloy, and this in turn has an influence on the mechanical and physical properties. The casting temperature has an important influence on the rate of cooling, and it is of importance to control the casting temperature within fairly narrow limits if uniformity of grain size is to be ensured in the

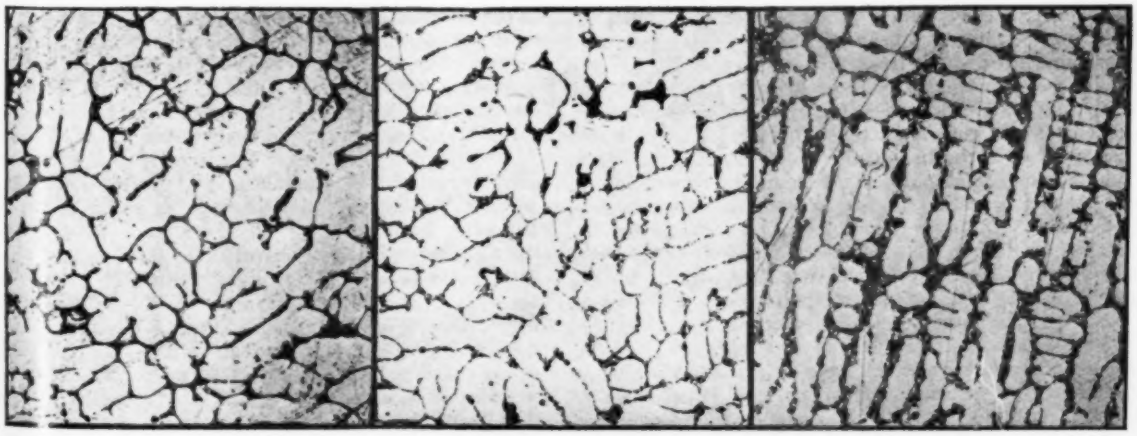


Fig. 7.—Structure of 15% silicon iron containing 0.30% carbon × 120.

Fig. 8.—Structure of 15% silicon iron containing 0.40% carbon ×120.

Fig. 9.—Structure of 15% silicon iron containing 0.50% carbon ×120.

Fig. 10.—Structure of 15% silicon iron containing 0.80% carbon × 120.

Fig. 11.—Structure of hypereutectic silicon iron, cast without "puddling." ×600.

Fig. 12.—A cast of the same iron as in Fig. 12 after "puddling" × 600.

castings. It has been found that if a high casting temperature is used there is a tendency towards the formation of big columnar crystals. On the other hand, a very low casting temperature produces a very fine grain indeed. At an intermediate casting temperature a moderate fine grain with soundness results. The optimum temperature range is between 1,220° to 1,280° C. (disappearing filament type pyrometer uncorrected), the precise temperature depending upon the size and cross-section of the casting.

The shrinkage on solidification of silicon iron is much greater than that of grey iron, and patterns are made to an allowance of $\frac{3}{16}$ in. per ft. The layout of moulds and cores has to be studied carefully to take care of this large contraction. Cores especially need care both in design and manufacture. They should be firm enough to withstand the pressure of the molten metal, and yet be sufficiently collapsible to give way under the contraction stresses. After the casting is completed the cores and the mould must be released, and the castings are stripped hot and transferred to an annealing furnace.

The Annealing of Silicon Iron

From a study of the binary thermal equilibrium diagram of the iron-silicon alloys, it would appear that for the acid-resisting irons in the region of 15% silicon no phase change takes place in the solid alloy. Although experimental work on the ternary system iron-carbon-silicon is not yet completed, all the work done so far shows that in the homogeneous alloys the only silicon phase is the alpha phase which would be expected from the binary diagram. In view of this, the exact purpose of annealing silicon iron castings has been a little obscure. It has been doubted by some that any useful purpose was served by such heat-treatment. Others, having experienced such things as spontaneous cracking in castings after normal cooling have considered some form of stress-relieving treatment as desirable.

Laboratory experiments have been made in order to examine the mechanical properties of silicon iron at annealing temperatures. The apparatus used consisted of a silica firebrick cut into the shape of a bridge on which a $\frac{1}{2}$ in. diameter bar of silicon iron could be strained centrally by a wire passing outside the furnace and supporting a hanging weight.

The bar was removed from the hot furnace at intervals and the deflection was measured. By using several bars from the same cast the deflection with time over a range of temperature was obtained. Deflection commenced at about 650° to 700° C., and was progressively greater at higher temperatures. The deflection, which at first increased rapidly, slowed up after a period of time. Probably the most important point brought out by the experiments was the lack of duplication of the results obtained. The points representing separate determinations were scattered widely around the mean curve, and two anomalies were found in the tests done at 700° C. when the bar bent in a direction opposite to the applied load. These curious results are regarded as evidence of the existence of internal stress in the "as-cast" material. The permanent deflection due to the stress applied at the annealing temperatures instead of being uniform in amount, in the direction of the applied stress, is increased or decreased by an amount due to the internal stress. Under certain circumstances the magnitude of this internal stress can be sufficient to bring about a negative deflection.

Experimental data shows that at temperatures in excess of 700° C. silicon iron shows a large degree of plasticity and toughness as compared with its properties at normal temperatures. At temperatures in excess of 800° C. excessive softening takes place which allows of easy distortion of castings. The toughness of silicon iron castings at these temperatures allows of their comparatively safe handling during stripping from the moulds.

Commercial silicon irons are found to contain small quantities of a carbide phase scattered at random through the whole cross-section of castings. The illustration (Fig. 13) shows a typical example of the occurrence of this phase which is unstable and easily decomposed on heat-treatment. The importance of the presence of this unstable phase in commercial castings on the possibility of cracking in service is recognised.

The adoption of an annealing treatment is a safeguard against the effects of internal stresses, and a treatment

which brings about the elimination of the unstable carbides under conditions which leave no undesirable stresses is an additional advantage. In the treatment used at the author's works the hot castings stripped from the moulds are transferred to an annealing furnace where they are maintained at a temperature of 800° C. for a period of four hours, followed by a period of slow cooling to room temperature. This operation is designed to allow the phase changes to go to completion at a temperature at which the metal is somewhat plastic and can adjust itself to internal strains without cracking. It is important not to exceed 800° C. or the metal becomes plastic and may deform under its own weight.

The Welding of Silicon Iron

Early trials with electric arc welding using 14/15% silicon iron electrodes proved disappointing, and the method adopted at the author's works is the oxy-acetylene method. In the production of certain castings, such as thermometer tubes, welding is very useful in that it allows of comparatively straightforward moulding methods. Long thermometer tubes, several feet long, closed at one end with core diameters of the order of $\frac{1}{2}$ to $\frac{3}{4}$ in. are conveniently cast as open-ended tubes, the closed end being formed afterwards by welding.

Standard oxy-acetylene welding equipment is used, and the procedure followed is as described below :—

Preheating.—The correct conditions of preheating are essential before commencing welding. It has been found necessary to ensure that the casting has reached a minimum temperature of 700° C. (medium red heat) before welding is commenced, and this temperature must be maintained at and around the weld until the completion of operations. Slow cooling to room temperature in an annealing furnace is necessary after the completion of the weld. Because of this need for preheating, soaking and annealing, it has been possible to weld very large castings with our present furnace arrangements.

Flux.—The flux found to be most satisfactory was one supplied by the British Oxygen Co. for the welding of cast iron, and it is believed to consist of equal parts powdered ferro-silicon, ferro-manganese and borax.

Flame.—Sound metal can only be deposited in the weld if a slightly oxidising flame is used. A neutral or reducing flame introduces hydrogen into the molten metal pool which is given out on solidification, and results in porosity. The jet size and gas pressures are carefully chosen, having regard to the size of the job, so that the flame is not too fierce. Silicon iron has a high fluidity at welding temperature and tends to run out of the weld unless carefully controlled.

Electrodes.—Sand-cast silicon iron electrodes of similar composition to the parent metal are used. Care is taken to remove sand by shot-blasting before welding.

Preparation of the Castings.—Assuming the weld to be made on a pipe, the ends of the joint are ground to a 45° "V" shape and a sand core is used to keep them in alignment during the early stages of the weld. A band around the pipe approximately 1½ in. from the joint is cleaned free from sand by grinding or shot blasting.

Method of Welding.—Welding is begun when the correct degree of preheat has been obtained, and the root of the "V" is welded first, taking care to ensure maximum penetration, and then the metal is added to fill up the space between the ends of the pipe. The pipe is rotated slowly, keeping the pool of molten metal always at the top centre until the whole weld is completed. Any irregularities are then smoothed off by sweeping movements with the torch.

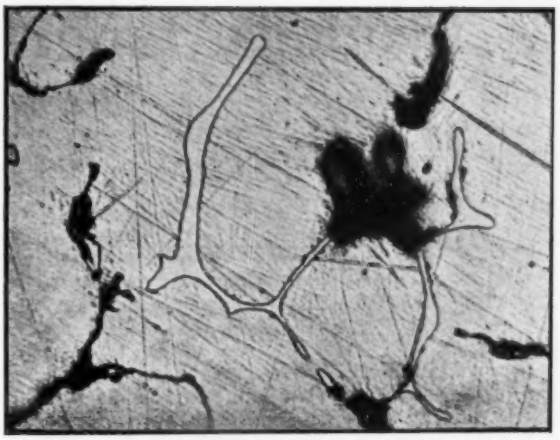


Fig. 13.—A typical example of the carbide phases in 15% silicon iron x 600.

Annealing.—When the weld is completed the casting should be maintained at 700°–800° C. for about an hour and then allowed to cool slowly to room temperature. Correct annealing is very important to prevent cracking troubles.

The Weldability of Cast Iron

(*Continued from page* 6).

cold draughts of air. Whenever possible provision for turning the part is arranged, so that welding is executed in a flat horizontal plane ; the fluidity of molten cast iron makes this almost imperative, if full control is to be held over the melt. A tip size slightly larger than that used for the same section in steel is used, as it is necessary to maintain a relatively large pool of metal during welding ; cast iron has also a greater capacity for heat than steel. A neutral or somewhat reducing atmosphere should cover the weld.

The filler metal generally used is a grey iron carrying an excess of silicon. It is more or less agreed that from 3–4% of this element is necessary to adjust discrepancies due to its volatilisation during welding, to effectively deoxidise, and also to ensure a workable deposit. Phosphorus is useful in promoting fluidity, but can be bad if present in amounts over 1%, mainly because of the effect of the phosphide eutectic upon ductility. Manganese and sulphur—both hardeners—are kept within low limits. Molybdenum, copper, chromium, nickel and titanium have been added to welding rods in small proportions with some advantage ; the general effect is to refine the structure. Plenty of first-class filler rods are available to-day. Small, well-distributed graphite, freedom from porosity, and surface unsoundness are features of these metallurgically controlled products (Fig. 14). Rods of a section varying from $\frac{1}{8}$ in. to $\frac{7}{16}$ in. cover most requirements, the thickness depending, of course, upon the type of application. It is a mistake to use a rod of extremely light section, as it may become burnt. On the other hand very thick rods can be clumsy, and in addition are apt to chill the melt. For welding alloy and white cast irons, and for special applications, rods are available of a suitably modified analysis.

(*To be continued.*)

Leaded-Manganese-Molybdenum Steel*

By T. Swinden, D.Met.

Central Research Department, The United Steel Companies Ltd.

Tests have been carried out on material from a non-leaded and a leaded ingot of manganese-molybdenum steel in order to determine the effect of lead (0·19%) on the mechanical properties, including the machinability; grain size and hardenability are also compared. The results are given in a slightly abridged form.

A COMPREHENSIVE study has been made of material from two ingots of manganese-molybdenum steel, to one of which lead was added, giving a content of 0·19% in the steel. The purpose of the tests was to obtain comparable data for the leaded and non-leaded steels.

ACID OPEN-HEARTH STEEL.—CAST ANALYSIS.

	%		%
Carbon	0·36	Phosphorus	0·045
Manganese	1·43	Molybdenum	0·43
Silicon	0·225	Nickel	0·16
Sulphur	0·032	Chromium	0·14

The ingots were cast uphill with brick tops; they measured 19 in. at the top, tapering to 17 in. at the bottom, and weighed 45 cwt. They were rolled by normal procedure into 4¼ in. by 2⅞ in. slabs, from which were subsequently rolled 4½ in. by ⅝ in. bars and 1 in. ¾ in. and ⅞ in. diameter bars.

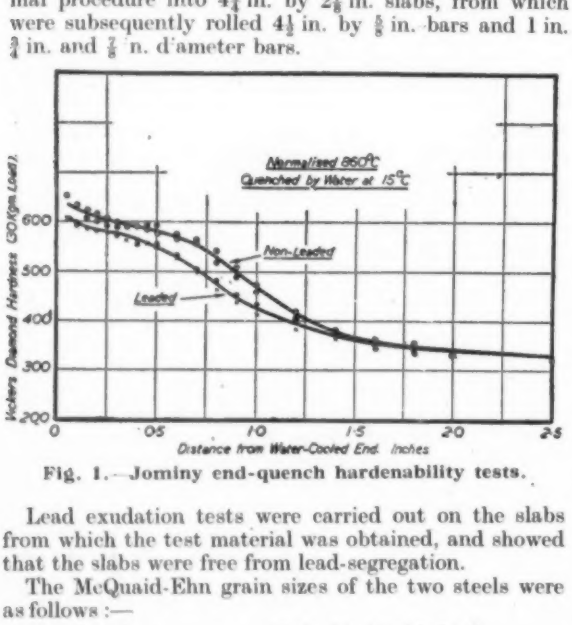


Fig. 1.—Jominy end-quench hardenability tests.

Lead exudation tests were carried out on the slabs from which the test material was obtained, and showed that the slabs were free from lead-segregation.

The McQuaid-Ehn grain sizes of the two steels were as follows:—

Leaded ..... Grades 4 to 6, mainly 5 to 6.
Non-leaded ..... Grades 6 to 7, mainly 6.

Standard Jominy end-quench hardenability tests were carried out on both steels. The results are shown graphically in Fig. 1.

On the basis of these results the non-leaded steel shows greater hardenability, as regards both maximum hardness and depth of hardening, than the leaded steel.

* Paper No. 13/1943 of the Alloy Steels Research Committee, presented at the recent Autumn Meeting of the Iron and Steel Institute.

Tensile and Impact Tests at Room

(*a*) *Tests on Flat Bars.*—Tensile and impact tests were carried out at room temperature on longitudinal and transverse specimens oil-quenched and water-quenched respectively, from 840° C. and tempered at 640°–650° C., the specimens being normalised at 910° C. for 2 hours before final treatment. The results are given in Table I.

TABLE I.

FLAT BARS, TENSILE AND IMPACT TESTS AT ROOM TEMPERATURE.

—	Longitudinal.		Transverse.	
	O.Q.	W.Q.	O.Q.	W.Q.
	Non-leaded.			
Max. stress, tons per sq. in.[a]	55·2	53·9	55·9	55·5
Yield-point, tons per sq. in.[a]	52·2	51·0	53·5	52·9
Elongation on $4\sqrt{A}$, %	25·3	26·7	16·0	18·0
Reduction of area, %	64·5	64·5	35·5	39·7
Izod impact value, ft.-lb.	72, 85, 63	94, 69, 85	13, 13, 13	12, 13, 12
Average	73	82	13	12
	Leaded.			
Max. stress, tons per sq. in.	53·8	55·2	56·1	55·0
Yield-point, tons per sq. in.	52·2	54·5	54·1	53·3
Elongation on $4\sqrt{A}$, %	25·3	24·0	16·6	17·3
Reduction of area, %	62·4	59·6	35·5	35·5
Izod impact value, ft.-lb.	74, 73, 73	66, 82, 58	11, 10, 10	11, 10, 11
Average	73	68	10	11

The difference in the effect on the tensile and impact properties of water-quenching compared with oil-quenching is negligible in both steels. The tensile-test results show only comparatively slight differences for the leaded compared with the non-leaded steel in both the longitudinal and the transverse directions. Both steels undergo a marked but normal reduction in elongation and reduction of area in the transverse compared with the longitudinal direction. The impact figures of the non-leaded steel are, in general, slightly superior to those of the leaded steel in the two directions. The figures obtained on transverse specimens are low in both steels.

(*b*) *Tests on Round Bars.*—Lengths of 1 in. diameter bar for tensile and impact tests were heat-treated to give 53, 58 and 65 tons per sq. in. tensile strength, the tests including determinations of the limit of proportionality and Young's modulus of each specimen. The results are given in Table II.

The maximum stress and limit of proportionality figures for the two steels treated to the three tensile strength values are respectively in very close agreement, whilst the yield-stress values differ only slightly, the general tendency being for those of the leaded steel to be lower than those of the non-leaded steel. There are only slight differences between the ductility figures of

TABLE II.

ROUND BARS. TENSILE AND IMPACT TESTS AT ROOM TEMPERATURE

Steel: Heat-treatment:	O.Q. 840° C. T. 660° C. 2 hours.	O.Q. 840° C., T. 650° C., 1 hour.	O.Q. 840° C., T. 620° C. 1 hour.
Non-leaded.			
Max. stress, tons per sq. in.	54·1	56·8	66·4
Yield stress, tons per sq. in.	52·0	52·0	64·0
Limit of proportionality, tons per sq. in.	48·4	50·0	60·0
Elongation on 2 in., %	25·0	26·0	21·5
Reduction of area, %	63·6	63·6	57·2
Izod impact value, ft.-lb.	79, 88, 79	81, 84, 72	58, 52, 58
Average	82	79	56
Modulus of elasticity, tons per sq. in.	12,800	12,750	12,800
Leaded.			
Max. stress, tons per sq. in.	54·2	56·8	65·6
Yield stress, tons per sq. in.	49·8	52·4	63·0
Limit of proportionality, tons per sq. in.	48·0	50·0	60·8
Elongation on 2 in., %	22·0	24·0	21·0
Reduction of area, %	61·6	61·6	57·2
Izod impact value, ft.-lb.	71, 59, 72	59, 65, 61	44, 44, 47
Average	67	62	45
Modulus of elasticity, tons per sq. in.	12,550	12,800	13,300

the two steels, as measured by elongation and reduction of area, the general tendency being for the leaded steel to be slightly less ductile than the non-leaded steel. The leaded steel shows appreciably less ductility, as measured by the notched-bar impact test, than the non-leaded steel when the tensile strength is within the range 54–66 tons per sq. in.

The modulus of elasticity figures of the non-leaded steel are similar, irrespective of tensile strength, but those of the leaded steel show a definite progressive increase in tensile strength from 54 to 66 tons per sq. in. When the tensile strength is of the order of 66 tons per sq. in. the modulus of elasticity of the leaded steel is appreciably greater than that of the non-leaded steel, but similar when the tensile strength is 56·8 tons per sq. in.

Tensile and Impact Tests at Elevated Temperatures

(*a*) *Tests on Flat Bars.*—Tensile tests were carried out at temperatures of 300°, 350°, 400°, and 450° C. on longitudinal and transverse specimens oil-quenched from 840° C. and tempered at 650° C., the specimens being normalised at 920° C. for 2 hours before final treatment. In addition, impact tests on the leaded steel were carried out on longitudinal and transverse specimens similarly heat-treated, to cover the range of temperature from 250° to 550° C., with a further test at 600° C. on a longitudinal specimen.

A standard rate of straining of $\frac{1}{24}$ in. per min. was adopted in the tensile tests, the rate bieng increased to $\frac{1}{4}$ in. per min. after passing the maximum stress. The results are given in Table III, and the elongation, reduction-of-area and Izod impact values are shown graphically in Fig. 2.

The maximum stress figures of the longitudinal and transverse specimens of the non-leaded and leaded steels are respectively similar, and all show a gradual decrease as the test temperature rises from 300° to 450° C. The elongation figures of the longitudinal and transverse specimens of both the non-leaded and leaded steels are at a minimum in the tests at 400° C. The values of the leaded steel are appreciably lower than those of the non-leaded steel, particularly in the transverse direction. The reduction of area values of the longitudinal and transverse specimens of the non-leaded steel increase as the temperature is increased from 300° to 450° C.; the leaded steel, however, shows minimum values at 400° C.,

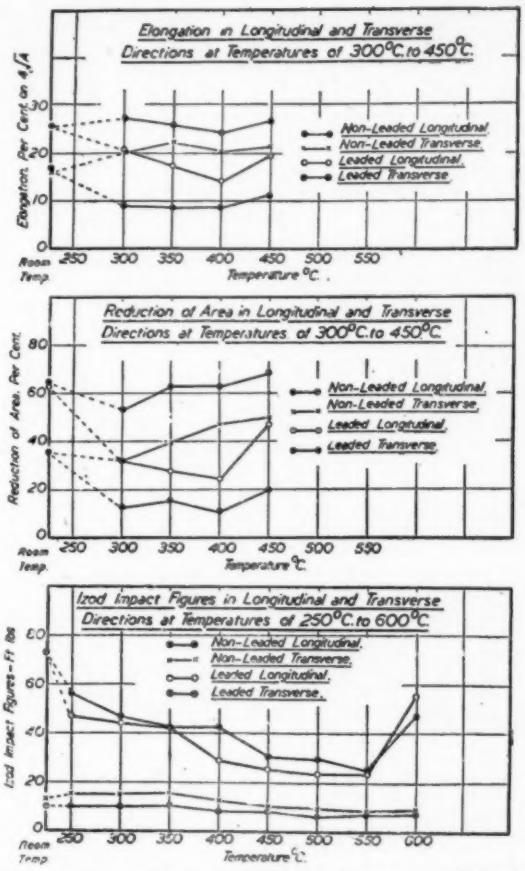

Elongation in Longitudinal and Transverse Directions at Temperatures of 300°C. to 450°C.
Elongation, Per Cent on 4√A
Non-Leaded Longitudinal
Non-Leaded Transverse
Leaded Longitudinal
Leaded Transverse
Room Temp
Temperature °C.
Reduction of Area in Longitudinal and Transverse Directions at Temperatures of 300°C. to 450°C.
Reduction of Area, Per Cent.
Izod Impact Figures in Longitudinal and Transverse Directions at Temperatures of 250°C. to 600°C.
Izod Impact Figures-Ft lbs

Fig. 2.—Elongation, reduction of area, and Izod impact figures compared for flat bar.

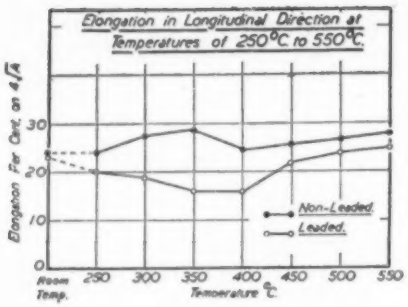
TABLE III.

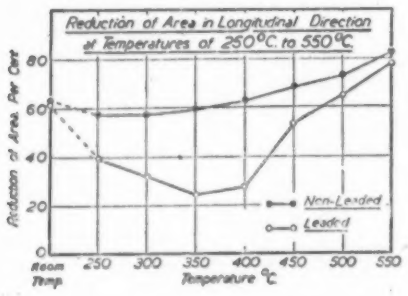
FLAT BARS, TENSILE AND IMPACT TESTS AT ELEVATED TEMPERATURES.

	Temp. of Test, °C.	Non-leaded.		Leaded.	
		Long.	Trans.	Long.	Trans.
Max. stress, tons per sq. in.[a]	300	58·1	57·5	57·9	57·2
	350	53·6	54·4	53·8	52·8
	400	50·1	49·6	50·2	49·4
	450	45·1	45·0	45·2	44·3
Elongation on $4\sqrt{A}$, %	300	27·3	20·0	20·7	9·0
	350	26·0	22·0	17·3	8·7
	400	25·3	20·3	14·0	8·7
	450	26·7	21·3	19·3	11·3
Reduction of area, %	300	53·2	31·9	31·9	12·6
	350	62·4	39·7	27·7	15·6
	400	62·4	46·8	24·1	10·6
	450	68·1	49·6	46·8	19·9
Izod impact, ft.-lb.	250	56	15	47	10
	300	46	15	44	10
	350	42	15·5	42	10
	400	42	12·5	29	8
	450	30	10	25	8
	500	29	9·5	23	6
	550	25	8	23	7
	600	47	9	55	7

in the tests on specimens taken in both directions. The figures for the leaded steel are markedly lower than those for the non-leaded steel, particularly in the tests on transverse specimens and especially at 400° C.

The results of Izod impact tests on the non-leaded and leaded steel in the longitudinal direction show decreasing values as the test temperature is increased





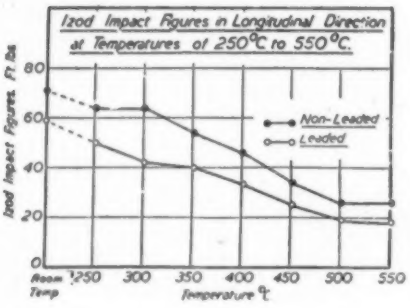


Fig. 3.—Elongation, reduction of area, and Izod impact figures compared for round bars.

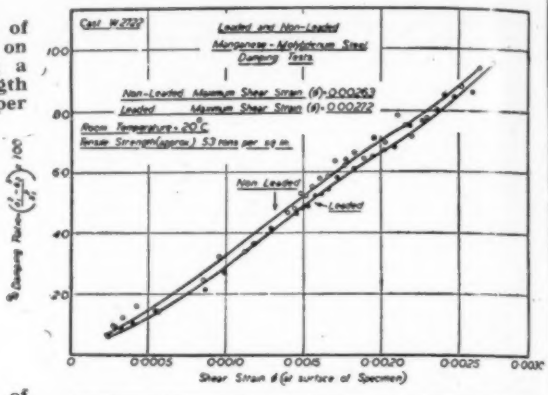

Cast W.2722
Leaded and Non-Leaded Manganese-Molybdenum Steel Damping Tests.
Non-Leaded. Maximum Shear Strain (ϑ)=0.00263
Leaded. Maximum Shear Strain (ϑ)=0.00272
Room Temperature=20°C.
Tensile Strength(approx.) 53 tons per sq.in.
Non Leaded
Leaded
% Damping Ratio
Shear Strain ϑ (at surface of Specimen)

Fig. 4.—Comparison of damping tests on steels having a tensile strength of 53 tons per sq. in.[2]

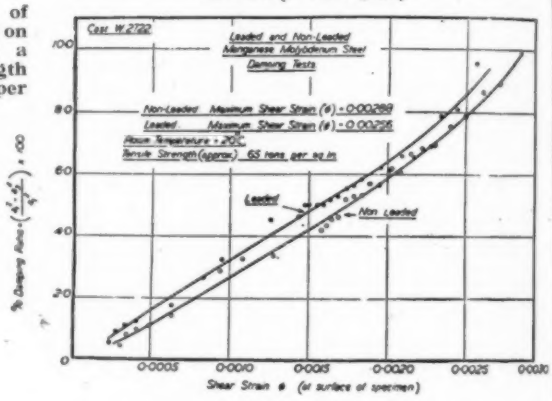

Cast W.2722
Leaded and Non-Leaded Manganese Molybdenum Steel Damping Tests
Non-Leaded. Maximum Shear Strain (ϑ)=0.00289
Leaded. Maximum Shear Strain (ϑ)=0.00295
Room Temperature=20°C.
Tensile Strength(approx.) 65 tons per sq.in.
Leaded
Non Leaded
% Damping Ratio
Shear Strain ϑ (at surface of specimen)

Fig. 5.—Comparison of damping tests on steels having a tensile strength of 65 tons per sq. in.[3]

from 250° to 500° or 550° C. and increasing values as the temperature is further increased to 600° C. The impact values of the leaded steel are generally slightly lower than those of the non-leaded steel up to 500° C., but the reverse is the case at 600° C. The values for the two steels are similar at about 24 ft.-lb. when tested at 550° C. The impact figures of transverse specimens of the non-leaded steel fall from 15 ft.-lb. at 250° C. to 8 ft.-lb. at 500°, and remain low up to 600° C. The impact values of the leaded steel are slightly lower than those of the non-leaded steel over the range of temperature from 250° to 600° C.

(*b*) *Tests on Round Bars.*—Lengths of $\frac{3}{4}$ in. diameter bar were oil-quenched from 840° C. and tempered at 650° C. Tensile and impact tests were carried out at temperatures of 250° to 550° C. in 50° intervals. The tensile specimens were strained at a rate of $\frac{1}{4}$ in. per min. up to the maximum stress, then at $\frac{1}{2}$ in. per min. to fracture. The results are recorded in Table IV.

There is no significant difference in the maximum stress figures between the non-leaded and leaded steels on testing up to 550° C.; both show a gradual decrease as the test temperature is increased from 250° to 550° C. The elongation, reduction of area, and Izod impact values are shown graphically in Fig. 3.

TABLE IV.
ROUND BARS. TENSILE AND IMPACT TESTS AT ELEVATED TEMPERATURES.

—	Temp. of Test, °C.	Non-leaded.	Leaded.
Maximum stress, tons per sq. in.[1]	250	57·6	57·8
	300	56·9	56·3
	350	52·3	53·7
	400	49·0	49·3
	450	43·2	44·3
	500	38·4	37·4
	550	32·1	34·1
Elongation on $4\sqrt{A}$, %	250	24·0	20·0
	300	27·3	18·7
	350	28·7	16·0
	400	24·7	16·0
	450	25·7	22·0
	500	26·8	24·0
	550	28·0	24·7
Reduction of area, %	250	56·7	39·7
	300	56·7	31·9
	350	59·6	24·1
	400	62·4	27·7
	450	68·1	53·2
	500	73·0	64·5
	550	82·3	78·0
		Av.	Av.
Izod impact values, ft.-lb.	250	67, 62 64	52, 49 50
	300	66, 63 64	44, 40 42
	350	56, 53 54	41, 39 40
	400	48, 45 46	33, 34 33
	450	37, 32 34	26, 25 25
	500	26, 27 26	19, 21 20
	550	25, 28 26	18, 19 18

Creep Tests

Barr-Bardgett creep tests were carried out at 500° C. on specimens machined from $\frac{5}{8}$ in. diameter bars, oil-quenched from 840° C. and tempered at 650° C. for 1 hour, having the following mechanical properties :—

	Non-leaded.	Leaded.
Maximum stress, tons per sq. in.	56·8	57·8
Yield stress, tons per sq. in.	52·0	52·4
Elongation in 2 in., %	26·0	24·0
Reduction of area, %	63·6	64·6
Izod impact value, ft.-lb.	81, 84, 72	59, 65, 61
Average	79	62
Barr-Bardgett creep stress value, tons per sq. in.	1·7	2·7

The leaded and non-leaded steels had identical Barr-Bardgett creep stress values at 500° C.

Damping Capacity

Measurements of damping capacity were made on a Cambridge torsional damping machine on specimens machined from bars heat-treated to approximately 53 and 65 tons per sq. in. respectively. In order to ensure that the stress applied in the damping tests did not exceed the limit of proportionality in torsion, determinations were made of the limit of proportionality in torsion, the following results being obtained :—

Steel:	Non-leaded.		Leaded.	
Heat-treatment:	O.Q. 840° C., T. 660° C. 2 hours.	O.Q. 840° C., T. 620° C. 1 hour.	O.Q. 840° C., T. 660° C. 2 hours	O.Q. 840° C., T. 620° C. 1 hour
Maximum stress tension, tons per sq. in.	52·7	66·2	52·8	65·2
Limit of proportionality, torsion, tons per sq. in.	25·4	29·0	26·1	31·0
Surface shear strain at limit of proportionality	0·00486	0·00557	0·00502	0·00605
Modulus of rigidity, tons per sq. in.	5,230	5,200	5,200	5,120

The results of the damping tests are shown in Figs. 4 and 5. It will be noted that the difference in damping capacity of the two steels is practically negligible when treated either to 53 or 65 tons per sq. in., although at the higher tensile strength the leaded steel has a slightly higher damping capacity than the non-leaded steel.

Fatigue Tests

Rotating cantilever (Wöhler) and alternating direct stress (Haigh) fatigue tests were carried out on both steels. Both plain and notched specimens were used in the tests on the Wöhler machine, but plain specimens only were tested in the Haigh machine. Tensile and impact specimens treated along with the fatigue specimens to tensile strength values of approximately 53 and 65 tons per sq. in. gave the results collected in Table V. In the original paper the results of the fatigue tests are recorded in an Appendix, and are shown graphically here in Figs. 6 to 9. A summary of the results is reproduced in Table VI.

In the rotating cantilever tests the plain and notched specimens of both steels treated to the higher tensile strength and the notched specimens treated to the lower tensile strength show similar values respectively for the fatigue ratios. The plain specimens of the non-leaded steel treated to the lower tensile strength have a slightly higher fatigue ratio than the corresponding specimens of the leaded steel.

The notch sensitivity of the higher tensile steels is similar, whilst at the lower tensile strength the non-leaded is slightly more notch-sensitive than the leaded steel.

The direct-stress fatigue-test results show a slightly higher fatigue ratio for the non-leaded than the leaded steel treated to the higher tensile strength, but the reverse is the case for the specimens treated to the lower tensile strength. The differences are, however, comparatively small.

Machinability

The principal object of this investigation is to study the properties other than machinability, although the first objective of adding lead is to improve machinability. A good deal of both laboratory and, what is more important, production data are available on this point, and, so far as the author is aware, no difference of opinion exists as to the improvement brought about in the machinability of this type of steel by the introduction

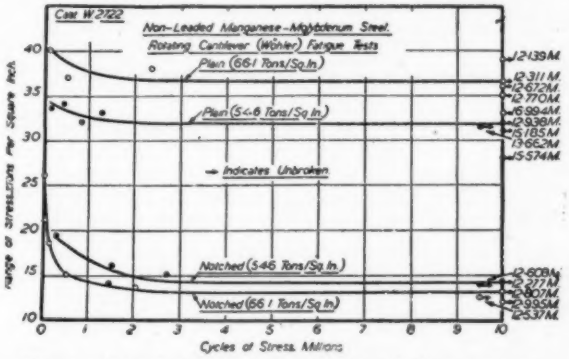


Fig. 6.—Rotating cantilever (Wohler) fatigue tests on non-leaded specimens, both plain and notched.

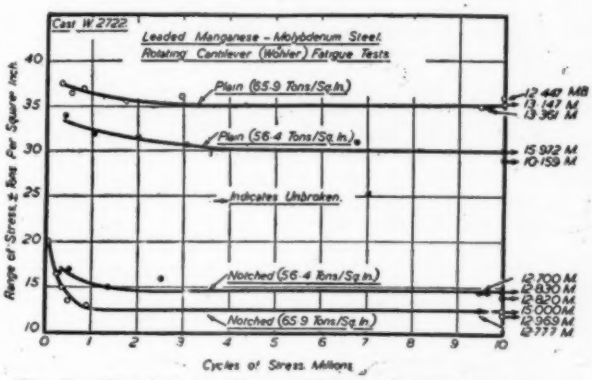


Fig. 7.—Rotating cantilever (Wohler) fatigue tests on leaded specimens, both plain and notched.

TABLE V.

ROUND BARS: MECHANICAL PROPERTIES OF MATERIAL USED FOR FATIGUE TESTS.

Steel:	Non-leaded.		Leaded.	
Test material:	¾ in. dia. Bar for (Wöhler) Rotating Cantilever Fatigue Test.			
Heat treatment:	O.Q. 840° C., T. 660° C. 2 Hours.	O.Q. 840° C., T. 620° C. 1 Hour.	O.Q. 840° C., T. 660° C. 2 Hours.	O.Q. 840° C., T. 620° C. 1 Hour.
Max. stress, tons per sq. in.	54·6	66·1	56·4	65·9
Yield stress, tons per sq. in.	52·7	61·8	54·0	60·1
Elongation on 2 in., %	24·5	20·5	24·0	19·0
Reduction of area, %	63·6	59·2	59·2	57·2
Izod impact value, ft.-lb.	67, 73, 79	54, 59, 50	54, 58, 58	48, 47, 46
Average	73	54	57	47
Test material:	¾ in. dia. Bar for (Haigh) Alternating Direct Stress Fatigue Test.			
Heat treatment:	O.Q. 840° C., T. 660° C. 2 Hours.	O.Q. 840° C., T. 620° C. 1 Hour.	O.Q. 840° C., T. 660° C. 2 Hours.	O.Q. 840° C., T. 620° C. 1 Hour.
Max. stress, tons per sq. in.	52·7	66·2	52·8	65·2
Yield stress, tons per sq. in.	50·2	63·6	47·1	62·0
Elongation on 2 in., %	26·5	21·0	24·0	20·5
Reduction of area, %	66·0	59·2	61·6	57·2
Izod impact value, ft.-lb.	90, 87, 81	51, 52, 47	72, 70, 74	41, 42, 42
Average	86	50	72	42

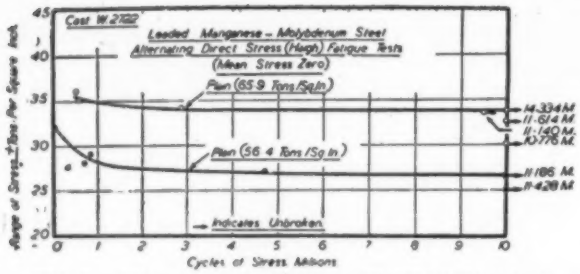


Fig. 8.—Alternating direct stress (Haigh) fatigue tests on leaded specimens, plain only.

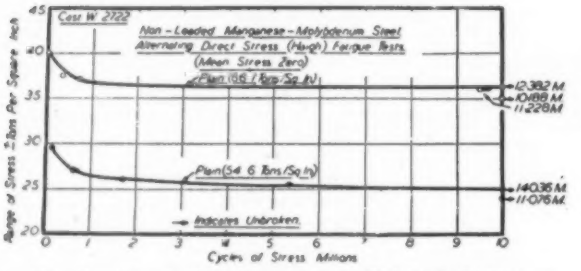


Fig. 9.—Alternating direct stress (Haigh) fatigue tests on non-leaded specimens, plain only.

of lead. One set of laboratory tests, which confirm improved machinability, is given by the following results from standard Losenhausen machinability tests on bars heat-treated respectively to 57 and 65 tons per sq. in. tensile strength.

Steel:	Non-leaded.		Leaded.	
Heat-treatment:	O.Q. 840° C., T. 650° C.	O.Q. 840° C., T. 590° C.	O.Q. 840° C., T. 650° C.	O.Q. 840° C., T. 590° C.
Brinell hardness	259	295	258	297
Machinability index	242	243	175	175

TABLE VI.

ROUND BARS: SUMMARY OF FATIGUE TESTS.

	Rotating Cantilever (Wöhler) Tests.			
Type of specimen:	Plain, P.	Notched, N.	Plain, P.	Notched, N.
Quality:	Non-leaded (LL).			
Maximum strength (MS), tons per sq. in.	54·64		66·08	
Limiting fatigue stress (LFS) ± tons per sq. in.	31·75	13·75	36·6	13·1
Fatigue ratio, MS/LFS	0·58	0·25	0·55	0·20
Notch sensitivity (P — N)/P	0·57		0·64	
Quality:	Leaded (KK).			
Maximum strength (MS), tons per sq. in.	56·44		65·92	
Limiting fatigue stress (LFS) ± tons per sq. in.	30·0	14·75	35·25	12·5
Fatigue ratio, MS/LFS	0·53	0·26	0·54	0·19
Notch sensitivity (P — N)/P	0·51		0·65	
	Alternating Direct Stress (Haigh) Tests. Mean Stress Zero.			
Quality:	Non-leaded.		Leaded.	
Type of specimen:	Plain, P.	Plain, P.	Plain, P.	Plain, P.
Maximum strength (MS), tons per sq. in.	52·68	66·24	52·84	65·2
Limiting fatigue stress (LFS) ± tons per sq. in.	25·0	36·25	26·5	33·75
Fatigue ratio, MS/LFS	0·47	0·55	0·50	0·52
Ratio of Haigh LFS to Wöhler LFS	0·79	0·995	0·89	0·96

The leaded steel has considerably better machinability than the non-leaded steel to the extent of 38% for the specimens of both higher and lower tensile strength the machinability indices being independent of the hardness in each case.

The Structure and Segregation of Two Ingots of Ingot Iron, One Containing Lead*

By L. Northcott, D.Sc., Ph.D., F.I.C., and D. McLean, B.Sc.

A CONSIDERABLE quantity of lead-bearing steel is now being used for purposes where ready machinability is required. Some information upon the mode of segregation of the lead in lead-bearing steel ingots is already available, but information was required upon the segregation of very low carbon content ingots of ingot-iron quality, the results of which are reported by the authors in their paper presented at the recent Autumn Meeting of the Iron and Steel Institute, an abridgement of which is given here.

Two 45-cwt. ingots, forming the fifth pair of a 75-ton melt, were cast; one being a plain ingot iron and the other a leaded ingot iron. The analysis of the pit sample was carbon 0·02%, manganese 0·037%, sulphur 0·039%, and phosphorus 0·01%.

The ingots were sectioned longitudinally, and the axial faces ground and polished. Sulphur prints were obtained from both the ingots and lead prints from the lead-bearing ingot. Three separate etching reagents were used to develop the macrostructure fully—namely, dilute nitric acid, acid ammonium, persulphate, and Stead's reagent. Different structures were obtained with each reagent.

Only in the case of the nitric-acid etching was the appearance of the ingot axial faces influenced by the presence of lead. The lead-bearing ingot had a bright uniform surface almost devoid of crystal structure, whereas the lead-free ingot showed the crystals very clearly and the core etched much darker than the rim.

Chemical Analysis.—Samples for chemical analysis were obtained from the twelve positions shown in Fig 1; the positions included those standardised by the Committee on the Heterogeneity of Steel Ingots to show

*A communication from the Armament Research Department, formerly the Research Department, Woolwich, to the Iron and Steel Institute, Sept. 1943. Advance copy.

heterogeneity. The results of the analysis are shown in Table I.

From top to bottom of the ingot the lead content decreases in the core, but increases in the rim. Apart from the higher lead content in the rim at the bottom of the ingot, the segregation of lead closely follows that of the segregating elements sulphur and phosphorus, in spite of the different modes in which these elements occur in steel and in the manner of addition of the lead. The carbon, silicon and manganese contents are so very low that segregation of these elements is virtually absent.

TABLE I.—CHEMICAL ANALYSIS.

Position.	Leaded Ingot Iron Ingot.							Ingot Iron Ingot.						Ingot Committees' Standard Position
	C, %.	Si, %.	S, %.	P, %.	Mn, %.	Pb, %.	Cu, %.	C, %.	Si, %	S, %.	P, %.	Mn, %.	Cu, %.	
1	0·02	<0·01	0·090	0·019	0·04	0·36	0·12	0·02	0·01	0·067	0·017	0·05	0·13	H
2	0·02	<0·01	0·072	0·016	0·035	0·32	0·13	0·02	0·01	0·058	0·015	0·04	0·13	D
3	0·02	<0·01	0·052	0·013	0·045	0·30	0·12	0·02	0·01	0·048	0·014	0·04	0·12	C
4	0·02	<0·01	0·049	0·013	0·05	0·26	0·12	0·02	0·01	0·032	0·011	0·04	0·11	B
5	0·02	<0·01	0·080	0·017	0·05	0·34	0·13	0·02	0·01	0·064	0·016	0·05	—	—
6	0·02	<0·01	0·053	0·014	0·04	0·29	—	0·02	0·01	0·051	0·014	0·04	—	—
7	0·02	<0·01	0·073	0·016	0·04	0·31	—	0·03	0·01	0·060	0·015	0·03	0·12	G
8	0·03	<0·01	0·054	0·015	0·05	0·22	0·11	0·05	0·01	0·037	0·014	0·04	0·12	F
9	0·02	<0·01	0·039	0·009	0·04	0·22	0·10	0·02	0·01	0·018	0·007	0·04	—	—
10	0·02	<0·01	0·043	0·009	0·04	0·24	—	0·02	0·01	0·024	0·009	0·05	—	—
11	0·04	<0·01	0·045	0·012	0·04	0·32	—	0·02	0·01	0·024	0·008	0·04	0·10	A
12	0·03	<0·01	0·042	0·011	0·04	0·71 0·53 0·66	0·63	0·02	0·01	0·026	0·011	0·05	—	—

Further data were obtained by sampling from a 1-in. wide slice cut across the middle of the ingot, ten samples being obtained from the edge to the centre line. These were analysed to determine the distribution in the direction of growth of the ingot. The results are given in Table II and plotted in Figs. 2 and 3. The determinations for carbon and sulphur were made by micro-chemical analysis, and it is probable that the values obtained are of higher accuracy than those given in Table I.

The general form of the curves is similar to that normally found in rimming steels, although the peak at the position corresponding with the commencement of the core is possibly more pronounced than usual, especially with respect to sulphur and carbon. The distribution of lead across the ingot follows the same general rule as for the other constituents examined, in spite of the different mode of occurrence of this element. One interesting feature brought out by the carbon analyses is that the carbon content at the extreme edge of the ingot was appreciably higher than that found in the core; as this feature occurs in both the leaded and the lead-free ingot, it is clearly not due to experimental error. A study has been made of the segregation in the steel ingots previously reported by the Ingots Committee, three examples of which having very similar carbon content. The results obtained with the present ingots show that the segregation found in one of the examples, which had been looked upon as an exceptional case, may indeed be quite representative of ingots of very low carbon content.

Microstructure.—Samples for micro-examination were obtained from the slices cut from the different positions selected for the tests on machinability. In view of the known difficulty with selective etching treatments for lead a comparison was made of the structures of the lead-bearing and lead-free steels. The structures were generally similar to those usually found in rimming steels. Numerous small inclusions of the sulphide type were observed in the core of the ingot, whereas in the rim the inclusions were for the most part too small to be identifiable. No evidence was obtained of indisputable lead deposits in samples from the cast ingot.

Samples which were hot-rolled contained short streaky particles in longitudinal sections of the leaded ingot, but not in the lead-free ingot; after annealing at 950° C. no difference could be observed in the leaded and lead-free sections, indicating a break-up or spheroidising of the supposedly lead streaks in the leaded ingot.

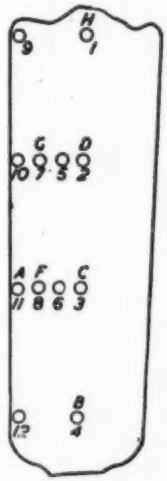


Fig. 1.—Positions from which samples were taken for chemical analysis.

TABLE II.

COMPOSITION FROM EDGE TO CENTRE AT MIDDLE OF INGOT.

Position.		Leaded Ingot.				Lead-free Ingot.		
		C, %.	S, %.	P, %.	Pb, %.	C, %.	S, %.	P, %.
Outside edge	1.....	0·05	0·041	0·014	0·32	0·045	0·034	0·016
	2.....	0·04	0·034	0·010	0·29	0·048	0·026	0·013
	3.....	0·04	0·020	0·009	0·21	0·034	0·019	0·014
	4.....	0·03	0·025	0·011	0·29	0·020	0·029	0·019
	5.....	0·04	0·039	0·013	0·36	0·032	0·054	0·021
	6.....	0·083	0·075	0·018	—	0·06	0·108	0·023
	7.....	0·035	0·066	0·017	0·30	—	0·072	0·027
	8.....	0·026	0·063	0·013	0·30	0·035	0·053	0·021
	9.....	0·032	0·045	0·016	0·31	0·023	0·053	0·023
Centre of ingot	10.....	0·051	0·043	0·013	0·24	0·025	0·074	0·019

Machinability.—As the purpose of lead additions is to improve machinability, tests were made of sections cut from six positions representing the core and rim at upper, middle and lower positions in each ingot. The upper position was about 9 in. from the extreme top of the ingot and the lower position about 1 ft. from the bottom. It is appreciated that the requirements for high machining speeds are usually confined to material in the forged and rolled condition and not in the condition as cast. The Oxford Airey machine used for the determination of relative machinability has been previously described.*

The figures given in Table III show the energy absorbed in removing one cubic inch of metal with a tool having an action similar to that of a milling cutter. The conditions of cutting were: Tool, high-speed steel; top rake, 20°; cooling agent, sulphonated oil solution; feed, 0·02 in.; depth, 0·02 in.

These figures clearly show that the energy absorbed in machining tests on the lead-bearing steel averaged 25% less than that with the lead-free steel. Furthermore,

* Kenneford. *Journal of the Institute of Metals*, 1939, **65**, p. 70.

in both ingots more energy was required in the machining of the outer rim than of the less pure core, the average figures being 3·58 and 2·91 ft.-lb. per cub. in. for the rim and core of the lead-free ingot, and 2·48 and 2·40 ft.-lb. per cub. in. for the lead-bearing ingot; in the latter steel a high lead content at the lower rim

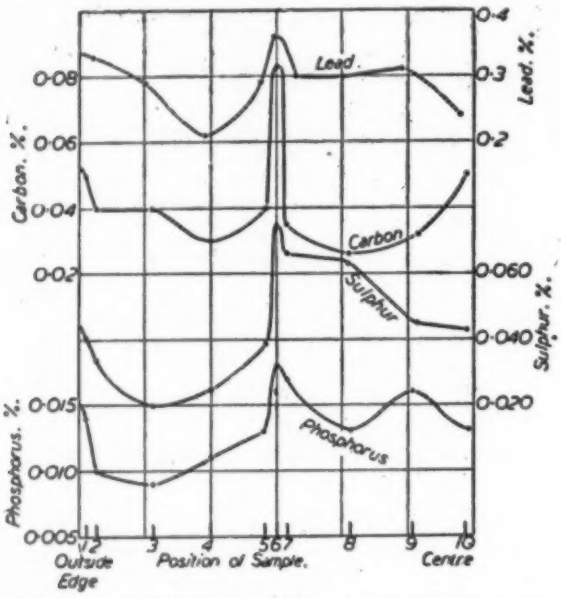


Fig. 2.—Lead-bearing ingot. Composition from edge to centre.

resulted in a lower energy figure at this position, and so modified the average for the lead-bearing series, but the presence of lead obviously had a notable effect in improving the machinability of the rim in the lead-bearing ingot.

Mechanical Properties.—Sections cut from the upper rim of each of the two ingots were hot-rolled to 0·3 in. and then annealed at 950° C. The mechanical properties obtained on the two materials were identical:

	Yield-point,* Tons per Sq. In.	Max. Stress, Tons per Sq. In.	Elongation, %.	Reduction of Area, %.	Vickers Diamond Hardness.
Lead-containing ingot iron	8·4	18·2	48	78	84
Lead-free ingot iron	8·3	18·4	43	77	85

* By dividers.

X-ray Examination.—With a view to detecting fine porosity and the presence of segregates of lead, an X-ray examination was carried out upon slices 1 in. wide and 0·25 in. taick cut from the centre to the edge of the ingot; samples were obtained from the top, middle and base of the ingot. For comparison purposes samples from the lead-free ingot were also examined. In order to determine the degree of local porosity a control sample was prepared by hot-rolling the slab cut from the rim of the lead-bearing ingot and machining this to the dimensions of the slices under examination, except that the thickness varied in ten steps from 0·25 in. to 0·025 in.; this control sample was radiographed alongside the cast sections. From a study of the X-ray negatives it appeared that:

(*a*) No evidence of lead was to be observed in sections cut from the rim of the ingots, except near the base where occasional large deposits were found associated with blow-holes.

(*b*) Several small globules of lead occurred in the core, more especially in the upper half of the ingot,

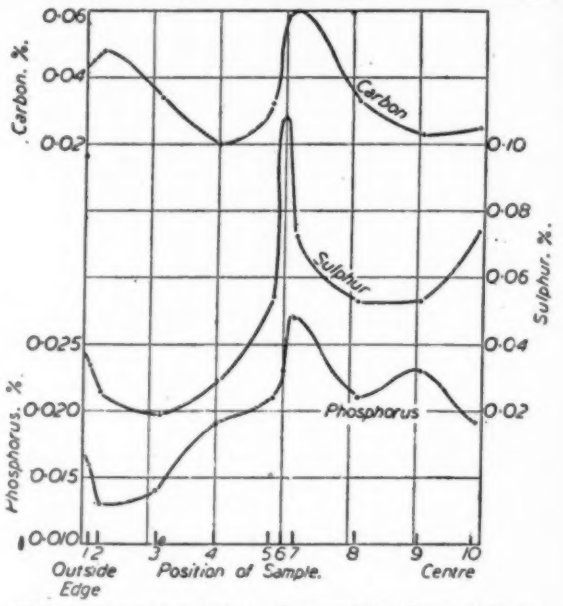


Fig. 3.—Lead-free ingot. Composition from edge to centre.

TABLE III.
RELATIVE MACHINABILITY.

Position in Ingot.	Energy Absorbed, Ft.-lb. per Cub. In.		Percentage Reduction in Energy Absorbed in Machining.
	Lead-bearing Steel.	Lead-free Steel.	
	× 10⁻⁴	× 10⁻⁴	
Upper rim	2·61	3·37	23·5
Upper core	2·45	2·83	13·4
Middle rim	2·45	3·74	34·5
Middle core	2·25	3·06	26·5
Lower rim	2·38	3·64	34·5
Lower core	2·51	2·83	11·3
Mean	2·44	3·24	24·7

particularly at the commencement of the core and also down the centre of the ingot.

(*c*) Correlation between the radiographs and the lead contents at different positions in the ingot suggests that X-ray examination, although valuable as showing the degree of lead dispersion, does not permit an estimate to be formed of the quantity of lead present.

(*d*) The core of the ingot appeared to be sound away from the large cavities, but the rim, which on visual examination gave the impression of being sound, was characterised by fine interdendritic columnar porosity. This porosity in the rim was also shown up by deep-etching polished sections, but it is thought that the radiographs illustrate the extent of the porosity much better than would etching. A comparison of radiographs of the rim with the radiograph of the forged and stepped bar showed that the interdendritic porosity may amount locally to as much as 10–20%.

Metallurgical Investigations of some Light Alloy Pistons from German Aircraft

A Summary of Investigations conducted by the Metallography Department of the Research Laboratories of High Duty Alloys Ltd.

under the direction of

C. Wilson, Ph.D., A.R.S.M., D.I.C.

General

THE following report contains a summary of the metallurgical data obtained by investigation of a number of light alloy pistons from German aircraft. The pistons were submitted by the Royal Aircraft Establishment, and the report is published under arrangements made by the Ministry of Aircraft Production.

These investigations and others which will be described in subsequent reports have been directed principally to determining the types of alloys in use and the general standard of quality adopted and to obtaining information relating to the methods of manufacture employed. The mechanical properties of each component have been determined, and the test-pieces were selected in relation to the directions of grain flow and the service to which the components have been subjected. Heat-treatment and general quality have been assessed by metallographic methods.

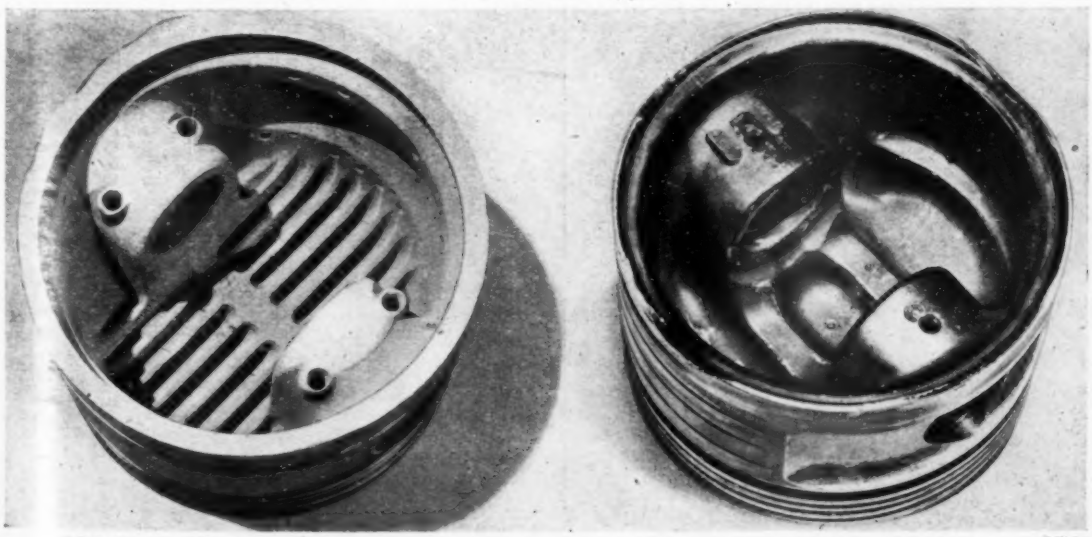
The investigations have been carried out with great care in order to obtain the maximum amount of information. Certain conclusions are drawn on the basis of experience gained with British Alloys and components, but no comparisons with such components are given.

Description of Samples

Seven pistons, representing four-engine types, have been examined, and an example of each type is illustrated in Fig. 1 to 4 inclusive. The pistons were similar in that each had been produced to close internal limits, with the object of reducing machining time. In each case, crown reinforcement had been obtained from ribs formed in the dies. The pistons had been utilised with very little internal machining and the finish produced by

Fig. 1.—Piston from Mercedes-Benz DB. 601N engine of Messerschmitt Me. 109 F2.

Fig. 3.—Piston of Bramo-Fafnir 323-P1 engine.



the forming dies was evidently considered to be satisfactory for duty under conditions of severe stress and high temperature.

Chemical Composition

With the exception of the piston from the Bramo-Fafnir engine, the material employed was of the eutectic silicon-aluminium type containing additions of copper, nickel and magnesium. The results of chemical analyses are given in Table I.

TABLE I.

Engine Type.	Cu, %	Ni, %	Mg, %	Fe, %	Si, %	Ti, %	Mn, %
B.M.W. 132 engine	0·88	0·91	1·08	0·53	12·04	0·09	0·04
B.M.W 801–A1 engine	1·08	0·99	1·15	0·56	11·42	0·09	0·0 2
Mercédès-Benz DB601N engine	1·05	0·95	1·17	0·42	11·4	0·09	0·02
Mercédès-Benz DB601 engine	1·09	1·15	1·06	0·71	11·65	0·12	0·02
Jumo 211 F1 engine	0·97	1·05	1·30	0·42	11·4	0·03	0·04
Jumo 211 A1 engine	1·02	0·93	1·16	0·43	11·7	0·08	0·06
Bramo-Fafnir 323-F1 engine	3·54	1·75	1·37	0·36	0·07	0·01	—

Mechanical Properties

The mechanical tests have been carried out with tensometer test-pieces obtained at selected positions. The test positions enumerated in the following table were selected to determine the effect of the operating temperatures on the properties of the material. The test results obtained at the base of the skirt or at the lower portion of the gudgeon-pin boss usually provide a close indication of the original properties and by comparison the gradual deterioration to the centre of the crown normally the hottest zone of the pistons, can be assessed.

TABLE II.
POSITION OF TEST-PIECES.

Position 1	Centre of crown.
Position 2	Gudgeon pin boss, normal to crown.
Position 3	Gudgeon pin boss, parallel with crown.
Position 4	Lower skirt, or lower gudgeon pin boss.

The results quoted in the above table represent the average of at least two tests. A further average of the results at position 4 indicates that the probable strength of the silicon-aluminium alloy was of the order of 21 tons per sq. in. ultimate tensile strength and with an elongation value of 3%.

TABLE III.
MECHANICAL TEST RESULTS.

Engine Type.	Position.	Yield-point, Tons/In.²	Ultimate Tensile Strength, Tons/In.²	Elongation % on 4/A.
B.M.W. 132 engine	1	10·4	14·1	7·0
	4	17·8	20·5	3·5
B.M.W. 801-A1 engine	1	11·5	15·2	5·5
	2	16·3	19·0	3·0
Mercédès-Benz D.B.601 N engine	1	8·9	12·8	9·0
	2	10·9	14·1	4·0
	3	12·6	*15·5	3·5
	4	16·0	19·5	2·0
Mercédès-Benz DB.601 engine	1	13·8	16·1	3·5
	3	16·9	18·3	2·8
Jumo 211 A1 engine	1	10·1	13·3	6·0
	3	—	18·5	2·5
	4	23·0	24·2	2·5
Jumo 211 F1 engine	1	11·0	13·7	3·5
	2	17·5	18·6	2·5
	3	15·3	17·3	3·5
	4	18·3	19·9	3·5
Bramo-Fafnir 323-F1 engines	1	13·0	19·4	11·7
	3	15·0	19·7	6·8
	4	23·5	26·7	8·0

*Approximates to 0·2% proof stress.

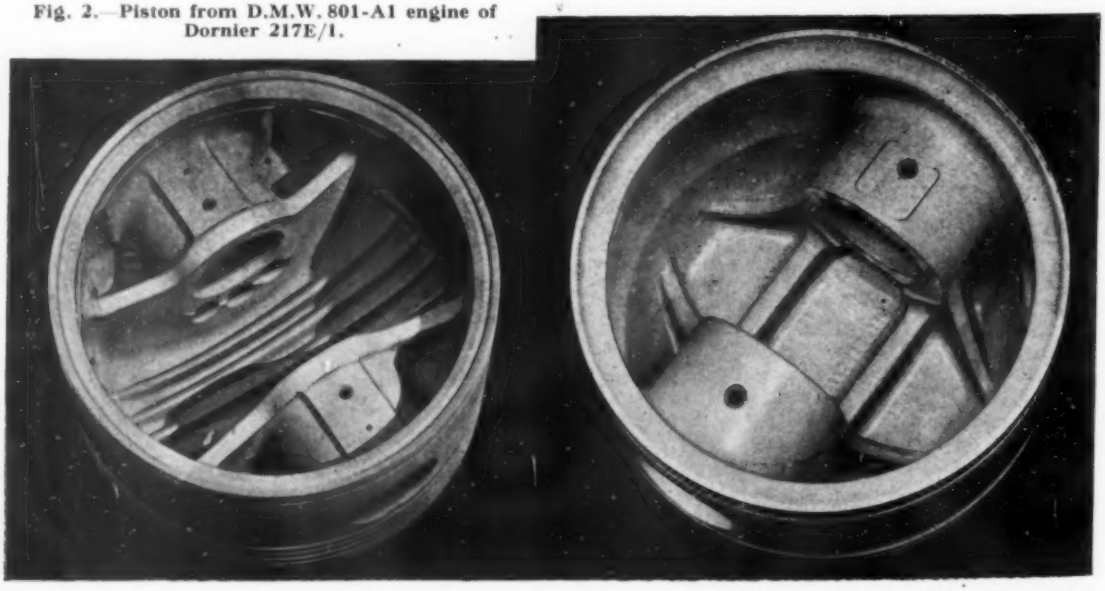
Hardness Tests

Considerable softening of the material takes place at the crown of a piston during service, and when hardness tests are carried out on a cross-section a marked hardness gradient between the crown and the skirt is observed. It is generally found that at the at the base of the skirt the material suffers only a slight deterioration in hardness and from this value an indication of the original hardness of the piston can be obtained. Values within the range 90–120 Brinell have been provided by the samples in the silicon-aluminium alloy.

By relating the hardness values obtained by a careful survey on a cross-section of a piston with the mechanical test results and the general structure, it is possible

Fig. 4.—Piston from Jumo 211 F1 engine of Junkers 88.

Fig. 2.—Piston from D.M.W. 801-A1 engine of Dornier 217E/1.



to obtain an indication of the operating temperatures. The isothermals contained in the sketch (Fig. 5) have been obtained in this way.

Macrostructure

The grain flow, as observed in etched sections prepared on a plane normal to the crown and through the gudgeon pin bosses, showed practically identical features in each case, and in the illustration the etched sections corresponding with Figs. 1 to 4 are shown in Figs. 6 to 9. At the centre of the crown, the flow lines were practically parallel with the external surface, but at the internal surface the metal in contact with the die had remained practically stationary. At the sides of the piston, the grain flow followed a direction practically normal to the crown, so that above the gudgeon pin bosses the flow lines emerged at an acute angle to the surface of the crown. The grain flow in the lower portion of the gudgeon pin bosses was practically parallel with the gudgeon pin axis.

Microstructure

In order to investigate the method of manufacture, the microstructure in planes normal and parallel with the crown has been examined. The methods of manufacture adopted for the four types of piston had evidently been very similar. In the specimens prepared parallel with the crown surface the original eutectic network remained largely undisturbed (see Figs. 10 and 12), whereas at right angles to this direction the microstructure showed fairly well-defined flow of the constituent particles. The eutectic areas remained, however, and merely showed elongation in the direction of flow (see Fig. 11).

The majority of the pistons had been manufactured in the eutectic silicon-aluminium alloy. In the samples examined, however, the fine modified structure normally associated with this alloy was not seen, and the influence of the modifying reaction seemed to have been largely dissipated prior to solidification.

The Bramo-Fafnir piston showed fine particle size (Fig 12).

Comments

The general features observed in the samples examined are those of pistons produced by pressing in dies, from extruded blanks, the close internal limits to which these pistons have been manufactured being typical of the press-forging method of production. The appearance of the microstructure is regarded as indicative of extruded stock, although the

Fig. 5.—Sketch showing estimated isothermals from B.M.W. 801-A1 engine.

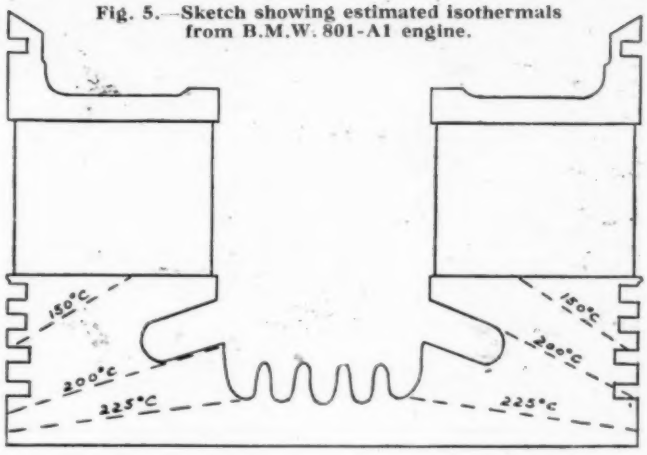


Fig. 6.—Macrostructure in section of Mercedes-Benz DB. 601 N piston.

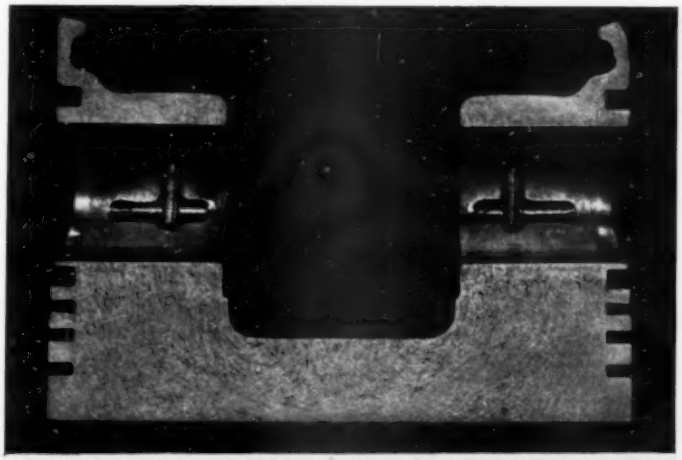
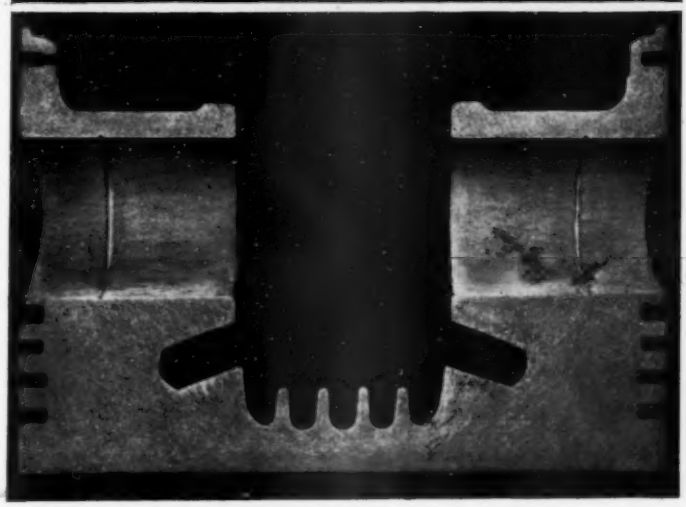


Fig. 7.—Macrostructure in section of B.M.W. 801-A1 piston.



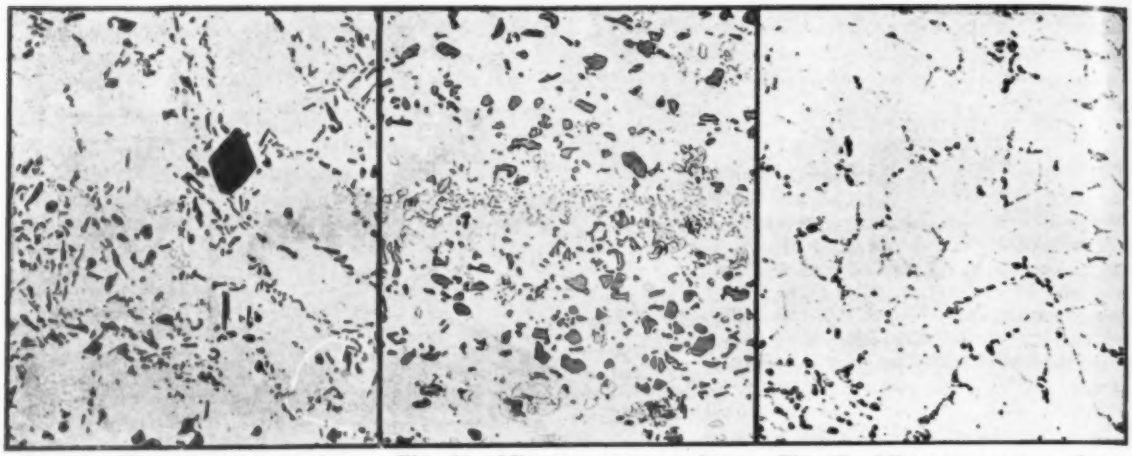


Fig. 10.—Microstructure on plane parallel with the crown surface of B.M.W. 801-A1 piston. × 200.

Fig. 11.—Microstructure on plane normal to the crown surface of B.M.W. 801-A1 piston. × 200.

Fig. 12.—Microstructure on plane parallel with the crown surface of Bramo-Fafnir 323-P1 piston. × 100.

Fig. 8.—Macrostructure in section of Bramo-Fafnir 323-P1 piston.

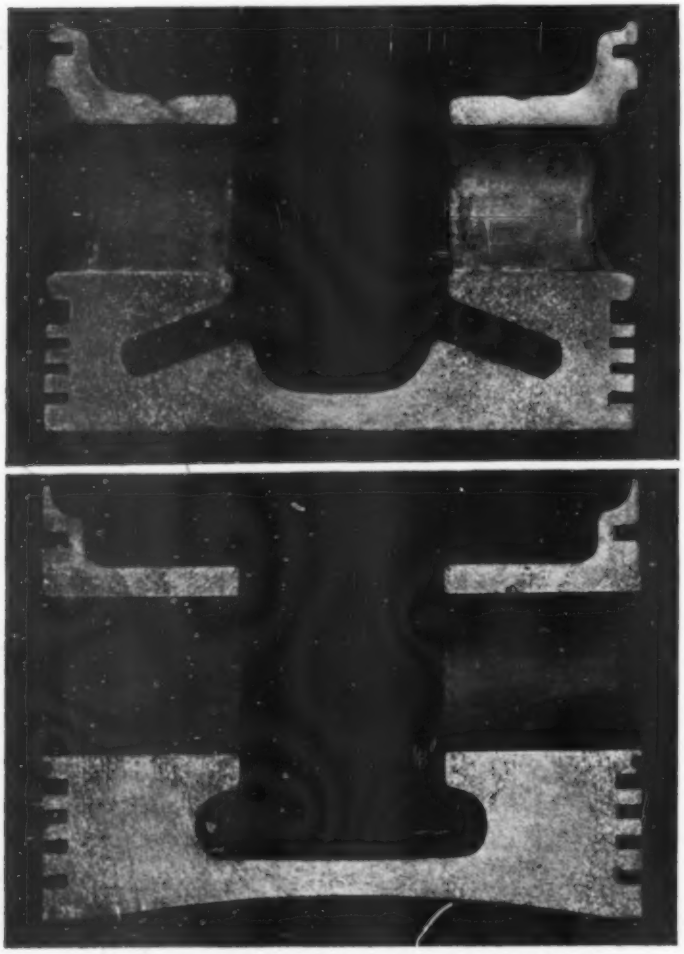


Fig. 9.—Macrostructure in section of Jumo 211 F1 piston.

degree of reduction in extrusion had not been great. It would appear that the Germans have adopted a standard manufacturing procedure and have practically standardised an alloy to be used for pistons. The procedure adopted constitutes a simple and economical mass-production method and provision for the variation in engine conditions seems to be made by ribs on the underside of the crown and by adjustment of the machined contours of the crown surface.

The pistons would be regarded as of good general quality, although the rather inferior microstructure of the B.M.W., Mercédès-Benz and Jumo pistons would not be expected to be accompanied by the best properties. It seems that the German manufacturers do not consider modification of the eutectic silicon-aluminium alloy to be important and evidently permit coarser particle size than is normally considered to be acceptable.

Aluminium Electrolytic Condensers

Robinson and Burnham* consider the electrolytic condenser in terms of its equivalent circuit, which consists of a condenser whose dielectric is aluminium oxide in series with a condenser whose dielectric is the solvent of the electrolyte. The oxide dielectric is shunted by a high resistance, and the solvent dielectric by the electrolyte resistivity. The characteristics of the condenser are functions of of temperature and frequency in terms of four variables are discussed, together with their chemical interactions.

* P. Robinson and J. Burnham. The Electrochemical Society. Preprint 83-24.

Electrolytic Production of Cobalt

A Review of Recent Work

Cobalt is an important constituent of many alloys now regarded as essential to metal-working tools, and especially to electrical engineering. For many purposes it is required in a very pure form—of greater purity than is possible by thermal reduction and subsequent refining. Recently, considerable attention has been given to electrolytic refining, and reference was made to the work of Oldach and Landau in our issue of August last. Other recent work on this subject is reviewed in the present article.

OWING to the difficulties encountered in the production of cobalt of more than 98% purity by means of thermal reduction and subsequent refining, considerable attention has been paid of late to the employment of the electrolytical method. Generally, it is known that the electrolytic production of cobalt should be carried out in an electrolyte as neutral as possible and of high Co concentration. The latter measure is important in so far as dilution leads to an increase in the hydrogen-ion concentration, and with it to increased hydrogen production, which reduces the yield and leads to spongy deposits of the metal. While references in literature are mostly concerned with general discussions of the possibility of the electrolytic production of the metal, C. Charmetant[1] relates experiences with the electrolysis of cobalt chloride with the employment of platinum electrodes in a water-alcohol mixture containing 450 g. ethyl-alcohol per litre. With a neutral electrolyte of 80 g. Co per litre and a current density of 200 amp. per sq. m., a current efficiency of 95% was obtained; but this efficiency decreases in time, and it also diminishes with increased current density, it is stated. According to Kigokada Nishihara,[2] the presence of small amounts of sulphuric acid greatly decreases the current efficiency. Various workers report that the electrolysis of solutions containing several metal-ions (iron, nickel, cobalt) leads to the production of alloys having a percentage of cobalt in excess of the average content of the solution.[3]

The possibility of producing cobalt sheet by the electrolysis of cobalt chloride at 70° C. and a current density of 100 amp. per sq. m. is briefly mentioned by K. Georgi,[4] admitting, however, that the sheet produced in this way is more brittle and harder than nickel sheet produced in similar manner. While most of the patents concerned[5] only mention the possibility of the electrolytic production of cobalt without giving any specific data, the French Patent 761,517 (1934) refers to an electrolyte containing 50–600 g. of $CoCl_2$ per litre, the current density varying from 200–1,500 amp. per sq. m. The addition of ammonium bifluoride amounting to 20% of the $CoCl_2$ contents is mentioned.

In their systematic investigations into the possibilities of electrolytic production of cobalt, P. Röntgen and K. Giesen[6] applied the data of the aforementioned patent to the electrolysis of a Co solution which was free of iron and copper, but contained Cl and So_4 ions, owing to the way in which the cobalt had been produced. Considerable difficulties were encountered, due to the production of hydrogen causing spongy metal deposits, but

1 C. R. Ac. Sci., Paris, **201**, 1935, p. 45.

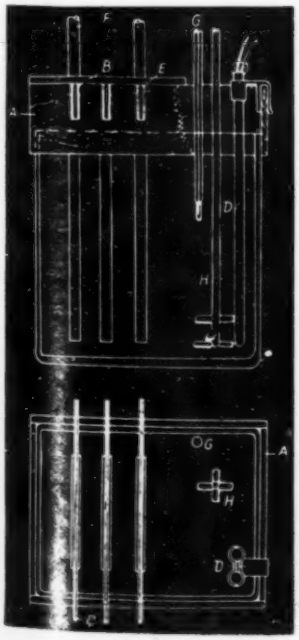


Fig. 1.—Electrolytic cell used by P. Roentgen and K. Giesen.

A. Wooden frame.
B. Slots for electrodes.
C. Electrode supports
D. Heater.
E. Wooden cover.
F. Vent.
G. Thermometer.
H. Stirrer.

2 See *Chem. Zentralbl.*, **109**, 1938, II, p. 366i.
3 R. Kremann. Elektr. Met. Kunde, in W. Guertler. *Metallographie*, vol. 2, Part I, H.2. C. R. F. Young and C. Egermann. *Met. Ind.*, **51**, 1937, p. 605. L. Weisberg. *Trans. Electroch. Soc.*, **73**, 1938, p. 435.
4 *Z. Elektrochem*, **39**, 1933, p. 210.
5 German Pat. 368, 610; 382, 226; U.S. Patent 1,055,653; 1,344,869; 1,336,765; 2,045,212.; Brit. Patent 416,158; French Patent 761517.
6 *Metall und Erz*, **33**, 1942, p. 7, 28.

Fig. 2.—Potential difference with temperature.

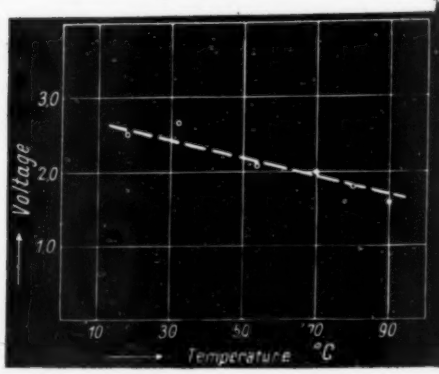
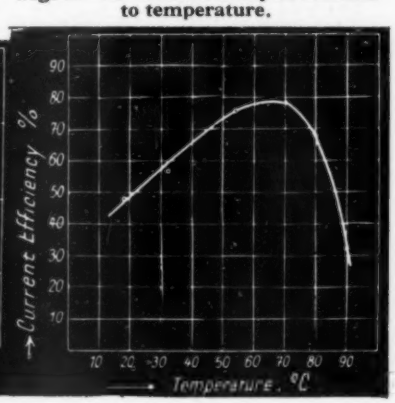


Fig. 3.—Current efficiency in relation to temperature.



these were partly overcome by increasing the concentration to 80 g. Co per litre. It was not, however, found possible to keep the nascent chlorine completely away from the cathode and to prevent chlorination of the deposited metal. The deposits obtained with a current density of 1,500 amp. per sq. m. were highly irregular and brittle, and scaled off as soon as a thickness of 0·1 mm. was reached. In these tests the electrolyte was held at 80° C., 1 mg. per litre of glue being used as colloidal agent. Also the initial current efficiency of 75–83% was found to decrease in time.

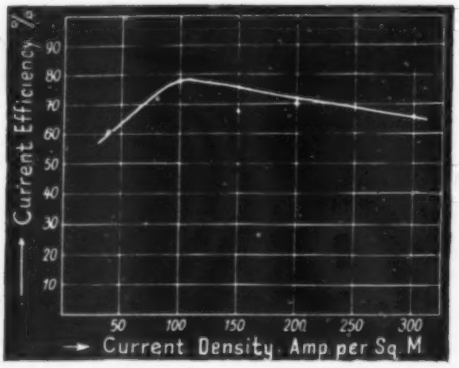


Fig. 4.—Influence of current density upon current efficiency.

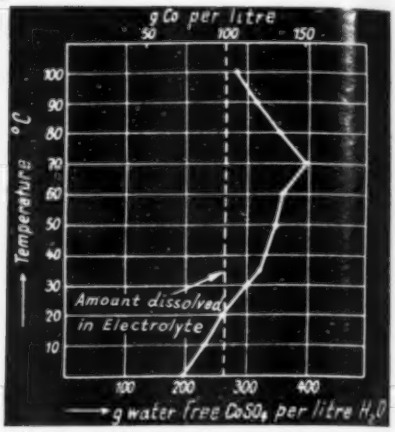


Fig. 5.—Influence of temperature upon solubility of $CoSO_4$ in water.

The Electrolysis of Cobalt Chloride

The electrolytic production of cobalt from cobalt chloride was investigated by P. Röntgen and K. Giesen[6] by preparing an aqueous solution of cobalt carbonate and hydrochloric acid, the carbonate having the following analysis :—

	%
Co	48
Cu	0·015
Fe	0·05
Mn	0·10
S	0·5
CaO	1·2

Ordinary glue was used as colloid. A glass tank of 3-litre capacity was used for the test (Fig. 1), the anode being a graphite plate of 100 × 100 × 10 mm. thickness, while the cathode was a sheet of nickel 100 × 100 × 5 mm. thickness in a wooden frame, in order to avoid high current densities along the edges. The results of the first test run, conducted with a current density of 100 amp. per sq. m. and the high bath temperature of 90° C. are given in Table I.

TABLE I.
INFLUENCE OF COLLOID PERCENTAGE.
Temperature : 90° C.
Electrolyte : $CoCl_2$: 100 g. Co per litre.
Current density : 100 amp. per sq. m.
Duration of test : 60 hours.
Distance between electrodes : 25 mm.

Colloid, Mg. per Litre, per Day.	Voltage, V.	Current Efficiency, %.	Deposit.
4	about 1·6	31·8	Dark, flaking off, strong formation of hydroxide.
6	about 1·57	29·9	Dark, flaking off, hydroxide formation.
10	about 1·6	34·6	Dark, porous, deposition generally somewhat improved, but brittle.

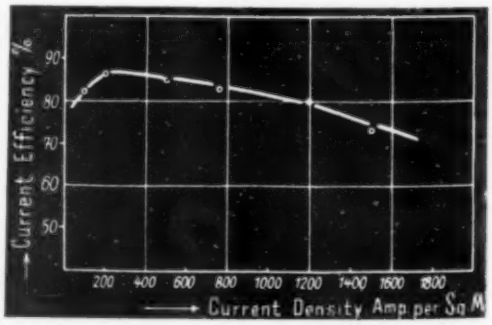
The inter-relationship between the potential difference between the electrodes and the temperature of the electrolyte is charted in Fig. 2, where the voltage is seen to decrease with the temperature. At high temperature the electro-deposition of the cobalt was found to be impeded by the considerable formation of hydroxide ; and this was particularly so in the case of low colloid additions. When the latter were increased, the surface appearance of the deposit improved noticeably ; but the deposit as such proved to be extremely brittle, and it tended to scale off the cathode during the test run. Also, the current efficiency is unsatisfactory. As this was assumed to be due to excessive temperature, another run with various temperatures was made, the results of which are given hereunder, to wit :—

TABLE II.
INFLUENCE OF BATH TEMPERATURE.
Electrolyte : $CoCl_2$, 100 g. Co per litre.
Colloid : 4 mg. per litre per day.
Current density : 100 amp. per sq. m.
Distance between electrodes : 25 mm.

Temperature, °C.	Current Efficiency, %.	Voltage, V.	Duration of Test Hours.	Deposit.
18	47·8	2·51	72	Formation of beads; deposit discontinuous.
32	56·3	2·66	72	Highly porous.
54	76·0	2·07	72	Formation of a few beads, and porous.
70	78·0	2·0	60	Smooth, easily removed.
80	67·5	1·82	72	Dark, some hydroxide.
90	31·8	1·60	60	Dark, flaking off, considerable formation of hydroxide.

From the temperature vs. current efficiency curve plotted in Fig. 3 it is seen that the current efficiency attains a maximum value between 60 to 70° C. Below 60° C. the current efficiency decreases slowly, but above 70° C. it drops sharply. The temperature of the electrolyte also exerts a marked influence upon the manner in which the formation of the deposit proceeds. While at room temperature only a few beads occurred, an

Fig. 6.—Current efficiency in relation to current density.



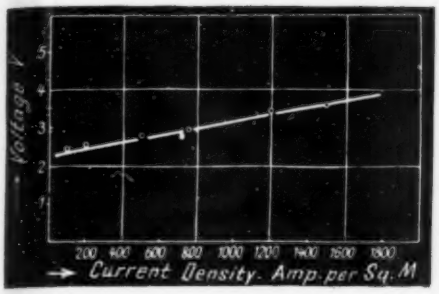


Fig. 7.—Current density in relation to voltage at 65° C.

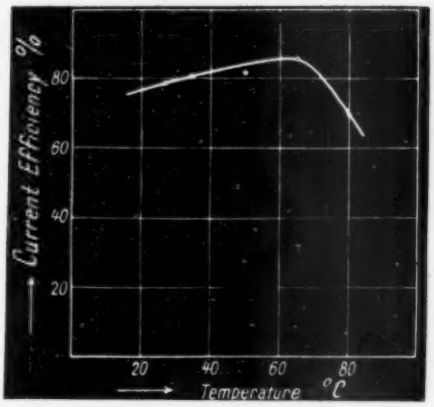


Fig. 8.—Influence of temperature upon current efficiency (with current density of 200 amp. per sq. m.)

increase in temperature to 32° C. resulted in a marked improvement of the surface of the deposit, and at 60° to 70° C.—that is, in the range of highest current efficiency, the tendency to the formation of beads was small. Another test run was made for the purpose of investigating the influence of the current density by varying the latter between 40 and 300 amp. per sq. m., the temperature of the electrolyte remaining constant at 70° C. This test yielded the following results, viz :—

TABLE III.

INFLUENCE OF CURRENT DENSITY.

Electrolyte : $CoCl_2$, 100 g. Co per litre.
Colloid : 4 mg. per litre per day.
Temperature : 70° C.
Duration of test : 40 hours.

Current Density, Amp./Sq. M.	Voltage, V.	Current Efficiency, %.	Deposit.
40	1·95	61·2	Porous, dark.
80	2·1	72·0	Porous.
100	2·0	78·0	Dense, smooth, easily removed.
150	2·15	68·2	Rough, dark, very poor.
200	2·3	70·0	A few beads.
300	4·5	66·8	Formation of beads, considerable amount of nascent hydrogen.

The influence exerted by the current density upon the current efficiency is evidenced by the curve shown in Fig. 4, which shows a maximum at 100 amp. per sq. m. In a further test run, the colloid percentage was varied, while the current density was maintained at 100 amp. per sq. m., the electrolyte having a constant temperature of 70° C. A minimum addition of 4 mg. per litre per day was found to be necessary for the production of a satisfactory deposit. Higher percentages proved worse than useless, as amounts of 6 or 8 mg. were found to produce brittle deposits covered with beads. No parcular influence of the colloid percentage upon the current efficiency could be observed.

Various amounts of boric acid were added in the final test run, in order to increase the conductivity of the electrolyte on the one hand and to produce a smoother and brighter deposit on the other. As Table IV, produced hereunder, shows, this result was obtained with an addition of 0·5% H_3BO_3.

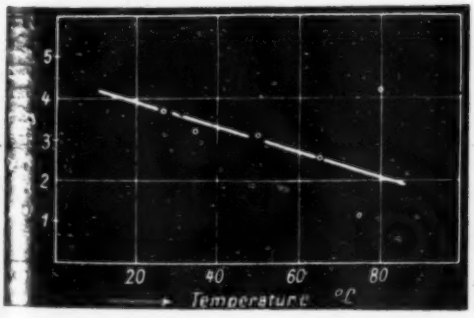
TABLE IV.

INFLUENCE OF VARYING BORIC ACID ADDITIONS.

Electrolyte : $CoCl_2$, 100 g. Co per litre.
Colloid : 4 mg. per litre per day.
Current density : 100 amp. per sq. m.

Boric Acid, %.	Voltage, V.	Current Efficiency, %.	Deposit.
0	2·0	78·0	Smooth, metallic.
0·25	2·0	72·0	Smooth, metallic.
0·5	2·25	73·0	Smooth, metallic, bright.
0·75	2·15	74·4	Formation of beads.

Fig. 9.—Voltage in relation to temperature for current density of 200 amp. per sq. m.



As Table IV shows, the highest current efficiency is obtained without the use of boric acid. Since brightness of the deposit is not required in the electrolytic production of cobalt, not much can therefore be gained by the addition of boric acid, particularly as the expected decrease in voltage did not take place.

The results of this investigation may be said to prove that the electrolytic production of metallic cobalt can be satisfactorily effected without the use of a diaphragm, the deposited metal having a purity of 99·91%. With an average voltage of 2 v. and a current efficiency of 78%, the power consumption per kg. of cobalt produced amounts to 2·34 k.w.-hour. In order to obtain optimum conditions, the following rules should be adhered to : Use of an electrolyte as neutral as possible, with a high degree of Co concentration ; current density of 100 amp. per sq. m. ; bath temperature of 60°–70° C. ; colloid addition of 4 mg. glue per litre per day. In order to eliminate obnoxious chlorine odours, a gas-tight cell should preferably be used.

The electrolytic production of cobalt from cobaltous sulphate has likewise been investigated by R. Röntgen and K. Giesen.[6] In these tests the concentration was also held at 100 g. cobalt per litre of electrolyte. As the solubility curve (Fig. 5) shows, this is the maximum concentration possible at 20° C. ; but it is considerably below the values obtainable in the range from 65°–75° C., which, as will be shown later, is the optimum range of operations. The test was conducted with lead anodes which proved fully serviceable, although they were found to be slightly attacked by nascent oxygen.

In the electrolysis of cobaltous sulphate, sulphuric acid is formed in the electrolyte, and it is therefore necessary to neutralise this by the periodic addition (about every 15 mins.) of an appropriate amount of cobalt carbonate, by weight approximating to twice the amount of the metallic cobalt precipitated :—

$$CoSO_4 \rightarrow Co + SO_4$$
$$SO_4 + H_2O \rightarrow H_2SO_4 + O$$
$$H_2SO_4 + CoCO_3 \rightarrow CoSO_4 + H_2O + CO_2.$$

There is, however, some difficulty in the determination of the acidity arising during the electrolytic process, as it has to be carried out by a somewhat complex potentiometric device. But in the test run under review, it was preferred to determine the amount of cobalt carbonate required on the basis of the electric current consumption of the cell.

A first test was conducted at 65° C. bath temperature with the current density ranging from 100 to 1,500 amp. per sq. m. The results obtained are listed below in Table V, and charted in Fig. 6.

TABLE V.
CURRENT EFFICIENCY AND VOLTAGE VERSUS CURRENT DENSITY.
Electrolyte : $CoSO_4$, 100 g. Co per litre.
Colloid : 4 mg. of glue per litre per day.
Duration of test : 24 hours.
Temperature of electrolyte, 65° C.
Distance between electrodes : 25 mm.

Current Density, Amp./Sq. M.	Voltage, V.	Current Efficiency, %.	Deposit.
100	2·4	81·6	Tending to form beads
200	2·6	86·0	Somewhat tending to form beads.
500	2·8	85·1	Brittle.
750	2·9	83·0	Brittle, flaking off.
1,200	3·4	80·1	Highly brittle, flaking off.
1,500	3·55	73·3	Highly brittle, flaking off.

The graph (Fig. 6) shows the current efficiency to rise with the current density until a maximum of 86% at about 200 amp. per sq. m. is reached, which is about 10% higher than the maximum efficiency reached in the aforementioned electrolysis of cobalt chloride. After that, the current efficiency falls slowly but steadily. The inter-relationship between the voltage and the current density at 65° C. is charted in Fig. 7, based on a distance of 25 mm. between the electrodes. A variation of the temperature of the electrolyte between 27° and 80° C. led to the following results :—

TABLE VI.
CURRENT EFFICIENCY AND VOLTAGE VERSUS TEMPERATURE.
Current density : 200 amp. per sq. m.
Colloid : 4 mg. of glue per litre per day.
Electrolyte : $CoSO_4$, 100 g. Co per litre.
Duration of test : 24 hours.
Distance between electrodes : 25 mm.

Temperature, °C.	Voltage, V.	Current Efficiency, %.	Deposit.
27	3·7	79·0	A few beads.
34	3·17	80·5	Porous, beads.
50	3·1	81·0	Irregular.
65	2·6	86·0	A few beads.
80	4·2	69·5	Highly brittle, flaking off.

The influence of the temperature of the electrolyte upon the current efficiency is shown by the graph (Fig. 8), indicating a maximum current efficiency at about 65° C.; while the influence of the temperature upon the potential difference between the electrodes is shown in Fig. 9, based on a current density of 200 amp. per sq. m. Generally, it was found that the surface condition of the deposited metal did not quite attain to that obtained with the chloride process. But this may be said to be fully in line with experiences made with analogous methods of nickel-plating, where a better surface appearance was obtained[7] with nickel-chloride solutions as compared to $NiSO_4$ solutions.

In order to ascertain whether the process could be improved by the employment of conducting salts such as sodium sulphate or boric acid, a number of tests were made with the addition of different percentages of either of these two salts. The results of these tests, made by P. Röntgen and K. Giesen,[6], are tabulated below :—

TABLE VII.
INFLUENCE OF CONDUCTING SALTS.
Electrolyte : $CoSO_4$, 100 g. Co per litre.
Temperature : 65° C.
Colloid : 4 mg. glue per litre per day.
Duration of test : 24 hours.

Conducting Salt.	Voltage, V.	Current Efficiency, %.	Deposit.
—	2·6	86·0	A few beads.
0·25% Na_2SO_4	2·5	90·9	Bright, a few beads.
0·5% Na_2SO_4	2·5	85·0	Dull, a few beads.
1·0% Na_2SO_4	2·6	87·1	Bright, a few beads.
5·0% Na_2SO_4	2·4	87·5	Bright, smooth, brittle, flaking off.
10·0% Na_2SO_4	2·5	90·4	Highly brittle.
5·0% Na_2SO_4	2·6	90·8	Bright, smooth, very brittle,
+ 0·25% H_3BO_3			flaking off.
5·0% Na_2SO_4	2·5	88·0	Bright, highly brittle, flaking off.
+ 0·5% H_3BO_3			
5·0% Na_2SO_4	2·5	88·1	Highly brittle.
+ 1·0% H_3BO_3			

The results indicate that not only does no improvement obtain with the use of conducting salts, but that the quality of the deposit is actually deteriorated by their addition to the electrolyte, the deposit flaking off as soon as a thickness of 0·3–0·5 mm. had been reached. Nor was it possible to decrease the potential difference between the electrodes by adding conducting salts. An excessive deterioration of the deposit was also found to occur with the presence of impurities, such as manganese and chlorine, in the cobalt carbonate. Highest purity of the electrolyte must therefore be considered as a prime requisite to the successful production of electrolytical cobalt.

A 70-hour test run with an electroplated cobalt cathode gave excellent results, the current efficiency amounting to 93·5% with a voltage of 2·5–2·6 v. This high efficiency must be ascribed to the fact that in this test an efficient stirring device was used to ensure the maintenance of a uniform concentration. The electrolyte used was a $CoSO_4$ solution with 100 g. Co per litre, to which 4 mg. glue per day were added as colloid.

Although the energy consumption of the cobalt-sulphate electrolytical process amounts to 2·48 k.w.-hour per kg. cobalt, as compared with 2·34 k.w.-hour required in the cobalt-chloride process, the former appears to be more attractice, since it avoids the formation of chlorine. In the commercial exploitation of the cobalt-sulphate process, the addition of the neutralising cobalt carbonate will preferably be carried out by a continually operating mechanical feeding device. Or neutralisation could be carried out external to the cell in a recirculating cycle.

7 W. A. Wesley and J. W. Carey. *Trans. Electrochem. Soc.*, 75, 1939, p. 209.

Palladium: Its History and Properties

By L. Sanderson

The discovery and development of palladium are briefly reviewed; various processes of extraction are briefly described, and some of its mechanical and physical properties are briefly discussed. Reference is made to some of its applications.

THE history of palladium is largely bound up with that of platinum, in whose ores it is found, though it is also found in the native condition and in Brazilian gold-bearing sand, associated with gold and silver. The first discoveries of its metallic crystals appear to have been in Brazil, but other deposits were found in the Antilles and the Urals. A different form of palladium was then located in the Hartz mountains, but investigation failed to confirm that this was an entirely separate and distinct form, and it is now regarded as the normal crystalline form.

Minerals in which palladium is found include *braggite* (Pt, Pd, Ni)S, whose palladium content is in the region of 20%; this mineral is one of the platinum group. British Guiana has a palladium-rhodium alloy in one of its river beds, this alloy being known as *potarite* (PdHg). *Clausthalite* contains mercury, palladium, and a number of the platinum metals, but has no rhodium. Additional elements in its composition include copper, silver, nickel and cobalt. In South Africa occurs *stibiopalladinite* (Pd_3Sb). Porpezite from Porpez in Brazil has 9·85% palladium in combination with gold and traces of iron and silver.

Table I shows at a glance the geographical location of the remaining palladium minerals, with the approximate percentage of the element in their composition.

TABLE I.

Location of Mineral.	Palladium, %.
U.S.S.R.	0·28–1·30
Colombia	0·35–1·66
California	0·6–1·95
Oregon, U.S.A.	0·15
Plattsburg, N.Y.	3·10
British Columbia	0·09–0·26
Spain	0·85
Australia	1·05–1·80
Borneo	1·41
Nevada, U.S.A.	0·2
Wyoming	?
Ontario, Canada	0·0068 oz. per ton

Numerous processes have been invented as isolating means for the recovery of the metal from ores of platinum. One of the earliest was that of Bunsen, who, in 1868, eliminated the majority of the platinum in the form of ammonium platinochloride, and precipitated the residual metals of the group by iron. The precipitate resulting was next heated with ammonium chloride and evaporated with fuming nitric acid. The residue was next taken up in water, and the metal precipitated in the form of potassium palladium chloride. The next stage was a purification process, which involved the dissolution of the chloride in hot water, followed by evaporation of the solution with oxalic acid, taking up the residue in potassium chloride, and filtering off any potassium platinochloride formed. Potassium palladium chloride is deposited by the filtrate, and when this was heated in a current of hydrogen, it left a palladium residue.

A more recent method is that of Gilchrist, who begins by extracting the metal from a palladium-bearing gold ore. This is first fused with potassium nitrate and silver, so that the gold equals approximately one-quarter of the entire mixture. The alloy is granulated, and afterwards digested with dilute nitric acid. The gold then becomes residual, and silver, palladium and copper are dissolved. The silver is precipitated out as a chloride, the palladium and copper being precipitated as a black powder by treating the acidic solution with zinc. The powder is then dissolved in nitric acid, and the resulting solution supersaturated with ammonia. By treating this solution with a slight excess of hydrochloric acid, yellow palladous diaminochloride is precipitated, which, when washed with cold water, may be ignited to yield palladium. The hydrochloric acid solution deposits the rest of the palladium, if any, and also the copper, when treated with iron.

This stage reached, Gilchrist then takes the crude palladium and dissolves it in aqua regia (300–500 vols. of HCl of sp. gr. 1·18, 75–100 vols. of water, and 60–70 vols. of nitric acid of sp. gr. 1·42). If any palladium oxide exists, it dissolves extremely slowly in the acid, so that it may be found in the undissolved residue. The solution is then evaporated in the steam bath to eliminate the greater part of the excess acid, and the syrupy liquor is dissolved in water and a small quantity of HCl. The solution is palladous chloride, which is diluted until it contains 40–50 grms. of palladium per litre. Should there be a considerable quantity of silver, it will largely separate out as silver chloride, the rest remaining in solution, and several reprecipitations may be needed to remove the entire quantity of silver. The silver chloride is filtered off, together with other residues, and ammonium hydroxide is added until no further precipitation of the flesh coloured compound ($Pd(NH_3)_4Cl_2PdCl_2$) takes place. Heating of the solution for 15–20 mins. in the steam bath follows. This redissolves the precipitate. The remaining steps are highly complex, and are best studied in one of the specialised papers. A good account appears in *Journal* of the American Chemical Society, 57, 2565, 1935.

Palladium itself is a metal of silvery hue, intermediate between silver and platinum. Its specific gravity lies between 11·3–11·8. Hardness is 4·8 on Moh's scale. Brinell hardness appears to be 52 for the cast metal and 49 for the annealed metal, though other workers give a figure of 61. Tensile strength of a palladium wire, 100 in. long and weighing 0·1987 grm. is 10 kgs. Elongation is 1·0% for the soft metal, and 24% for annealed metal, in 2 in. Coefficient of thermal expansion is 0·041180 at 20° and 0·041176 at 40°. Specific heat is 0·05928 between 0–100°. Melting-point appears to lie between 1,460°–1,500° C. This shows that it is one of the most easily fused of all the platinum group of metals. Its electrical conductivity is 13·97 as compared with silver = 100.

Ultimate tensile strength of palladium is 47,000 lb. per sq. in. in the hard condition, and 20,000 lb. in the

annealed condition. Erichsen number is 7·6 hard, and 12·0 annealed. Resistivity in ohm./mil. ft. at 0° C. is 61. Temperature coefficient of resistance 0–100° C. is 0·0038. Being inherently a soft metal, it is hardenable only by the addition to it of other elements. It is widely employed in dental work, in jewellery, and as a material for arcing contact points. Another purpose is as a catalyst, while it has been employed as a type of foil for various ornamental effects, and as a photographic printing paper. Additional uses are for astronomical instruments and for the construction of springs and movements in clocks.

It is often alloyed with platinum, mainly as a means of lessening cost, since it is cheaper and lighter than any of the other platinum metals.

The atomic weight of the metal is 106·7. Density is elusive, but 20° C. is usually taken as 12·0 g. per c.c.

Palladium is also sometimes alloyed with gold. The result is a highly ductile alloy with a tensile strength of 50,000 lb. per sq. in. in the annealed condition. It does not markedly lower the melting point of gold, as do other alloys. Moreover, it lightens the colour of the more precious metal, and greatly improves the corrosion resistance. For this reason alloys of 70–80% gold, with palladium or gold-platinum-palladium alloys, have been used for laboratory and other crucibles and for chemical plant, including rayon spinnerets. An 80–85% gold-palladium alloy resembles in mechanical properties a 22-carat gold, and is valuable in dental work, where it is desired to secure a white, readily workable sheet metal. Silver is also improved in resistance to tarnishing by the addition of palladium, and it is this ability to increase stain-resistance that makes the metal so useful in jewellery and dental work.

The palladium-platinum alloys range from 10 to 75% of palladium. A 10% palladium-platinum alloy has a Brinell hardness of 145 in the hard state and 80 in the annealed state. Ultimate tensile strength is 45,000 lb. per sq. in., Erichsen No. 11·5, and resistivity ohms./mil. ft. at 20° C. 120. Temperature coefficient of resistance 0–160° C. is 0·0014. A 75% palladium-platinum alloy has the figures 150, 80, 43,000, 11·5, 130°, and 0·0014 respectively.

Nickel hardens palladium very quickly.

Native palladium is dimorphous. It is soluble in nitric acid, particularly if the acid contains oxides of nitrogen. In the finely divided condition, it is to some extent soluble in hydrochloric acid. It oxidises when fused with caustic alkalis. It combines with fluorine and chlorine at a dull red heat, but not with iodine, while there is virtually no reaction between palladium and bromine. It combines directly with sulphur.

The atomic number of the metal is 46. It appears to lose 0·6% of its weight when heated in air for 30 hours at 1,300° C. At this temperature, its volatility in air is three times that of platinum. It is obtainable by electrodeposition, a bright deposit being produced by the employment of a current density of 0·05 amp. and 1·2 volts. The deposition is from a solution closely akin to those used in the electrodeposition of platinum. An ammoniacal solution of palladous diaminochloride appears to yield satisfactory results. The metal is capable of being obtained in various physical states—e.g., the compact metal, in a crystalline form, as a spongy metal, as palladium black, and as colloidal palladium.

Palladium black is obtained by precipitating from a solution of palladium salts by a reducing agent, such as sodium formate. Palladium black embodies 1·65% oxygen that can be eliminated only by ignition in a current of hydrogen. When dried at 100° C., it contains 0·72% water, so that if the 138 vols. of oxygen is present in the form of palladous oxide, the black will contain 86·59% palladium, palladium oxide 12·69, and water 0·72%.

In palladised asbestos, the finely divided palladium is spread over asbestos, when it provides an extremely effective palladium catalyst.

Not many figures of production are available, most countries including their output with the platinum figures, but in 1940 the United States produced 14,773 metric tons of the metal, a considerable increase over 1939. 4,564 tons were recovered by refiners.

Causes of High Dewpoint Temperatures in Boiler Flue Gases

IT is widely recognised that the serious difficulties brought about by acid deposits in the air heaters and economisers of certain modern boiler plants fired by mechanical stokers are due to abnormally high dewpoint temperatures of the flue gases in such installations, and that the elevation of dewpoint is a result of the gases containing an excessive amount of sulphur trioxide.

The existence of such conditions was pointed out by Johnstone* as a result of investigations, but the primary reason for the exceptional conditions was not determined. Despite a great deal of research work which has since been devoted to the problem, no satisfactory explanation that could be supported by general field experience has been forthcoming.

Numerous conditions, including the manner of burning the fuel, steam leakage into flue gases, high moisture content of the fuel, the design and construction of combustion chambers, the absence or presence of secondary air and the nature of the particles carried in the gas stream, have all in turn been suggested as possible causes. Although each of these factors may have some bearing on the matter, it has not been possible to show that any of them are primarily responsible.

As a result of a series of experiments described by W. F. Harlow, in a recent paper submitted for discussion at a Joint Meeting of the Institution of Mechanical Engineers and the Institution of Electrical Engineers, evidence is submitted to show that high dewpoint can be caused by the increased temperature of the superheater tubes, the oxide film catalysing the conversion of the sulphur dioxide formed by combustion into sulphur trioxide, thereby increasing the sulphuric acid content of the gases. The trouble is confined chiefly to plants fired by mechanical stokers, pulverised fuel plants generally being unaffected.

It is suggested that by coating the catalytic surfaces with lime or fine pulverised fuel ash the action is prevented and, by spraying these materials on the tubes at intervals, the formation of acid with its serious consequences may be prevented.

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*H. F. Johnstone. "An Electrical Method for the Determination of the Dewpoint of Flue Gases." University of Illinois Engineering Experiment Station, Circular No. 20, 1929.

METALLURGICAL DIGEST

SINCE the inception of this Journal we have made a practice of publishing an occasional digest of important work presented in the technical press, but no effort has been made to preserve continuity. The usefulness of these digests has, on many occasions, been commented upon by readers, and in order that we may be of greater service in this respect, we are commencing under the above title a new feature in which, for the time being, we shall endeavour to publish digests of foreign literature. Every effort will be made to ensure that the digests are reliable and that they will be maintained on a high level.—ED.

The Cleaning of Heated Steel before and during Rolling and Forging

By Gunnar Wallguist

IN the heating of steel for rolling and forging, surface oxidation and various other forms of surface contamination occur. Also oxidation occurs during the actual working process. If this contamination is not removed, it may become worked into the surface. In order to avoid this, the surface layer must be removed before, and also even during the working operation. This can be done by (*a*) mechanical means, (*b*) steam jet, (*c*) high-pressure water jet. Of these, the last-named is the latest and most effective. The cleaning action is produced partly by chilling the oxide film, which is thus loosened, and partly by the force with which the water particles strike the surface.

This method originated in the U.S.A. and was first used in Sweden in 1935. It was studied in detail by the author from 1937 onwards.

For effective cleaning, it is desirable to use plane jets from a row of nozzles. The force required depends on the nature of the surface layer and the form of the article. The total force from a nozzle can be written as:—

$$P_{total} = k_1 Q \quad \ldots\ldots \quad (1)$$

where P is in kg., Q is the quantity of water issuing from the nozzle in litres per minute, k depends on the water pressure and on the form of the nozzle. With a plane jet, the dependence of k on the water pressure is shown in Fig. 1. The factor Q is given by:

$$Q = 0{\cdot}84.\eta.A.p \quad \ldots\ldots \quad (2)$$

where A is the area of the nozzle, p the water pressure in kg./cm² and η is a nozzle factor. This varies from 0·52–0·62 for the last of the nozzles illustrated in Fig. 2, which shows four widely used types. The first two incorporate filters. The nozzles are placed according to Fig. 3. The angle of inclination with the normal to the sprayed surface (β) is about 15°; γ, the angle of obliquity of the planes of the jets with the cross-section of the material can be 15°–45°. The breadth of the nozzle opening (*b*) is 10–12 mm. and the angle (α) of the spray is 30°.

As a measure of the degree of purification of the surface, the percentage of oxide skin removed may be used, but this is not quite good enough, since it does not take into account the distribution of oxide over the surface. If there are isolated patches or streaks the degree of purification may be quite high, but the final product might still be unusable.

Experiments were made, rolling to thin sheet, and to bar, wire and strip. Measurements of the degree of purification were combined with microscopic examination and practical tests. The pressures of water used were similar to those employed in modern installations. These usually vary from 70–100 kg./cm², normally closer to the upper limit than the lower; they are produced by means of centrifugal pumps, working with or without an accumulator. The size of the pumps varies from an input of 30–500 litres per min.

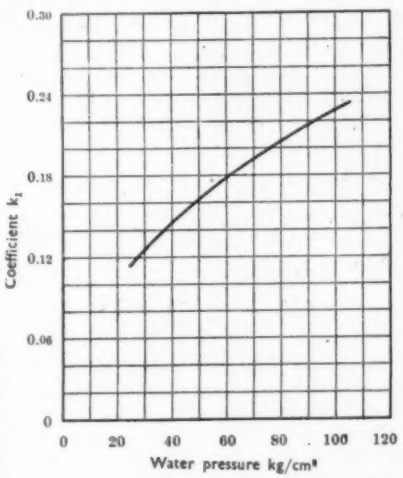

Fig. 1.—Relation between k_1 and water pressure.

The cleaning operation was carried out after heating and before rolling, sometimes with an additional cleaning between rolling passes.

Experiments carried out on mild steel containing 0·1% carbon showed a large improvement in surface purity in rolling to thin sheet, average figures being 330 grams/metre² initially, 80 grams finally.

From *Jernkontorets Annaler*, 1943, vol. 127, No. 3.

In rolling to bar the amount of oxide actually rolled into the surface depends on the type of alloy, the manner of rolling, and degree of reduction. With larger percentages of alloying additions, the degree of oxidation is usually greater, and light rolling passes and small amounts of total reduction result in a poor surface, owing to the effect of mechanical cleaning produced by the actual rolling process. The water-jet method was shown to be very suitable for use in the rolling of stainless nickel-chrome steels and hollow drill steel.

In rolling to wire the total degree of reduction is large and the mechanical cleaning effect is therefore considerable.

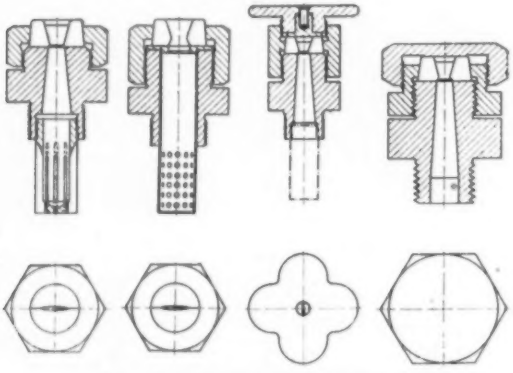

Fig. 2.—Some typical nozzles.

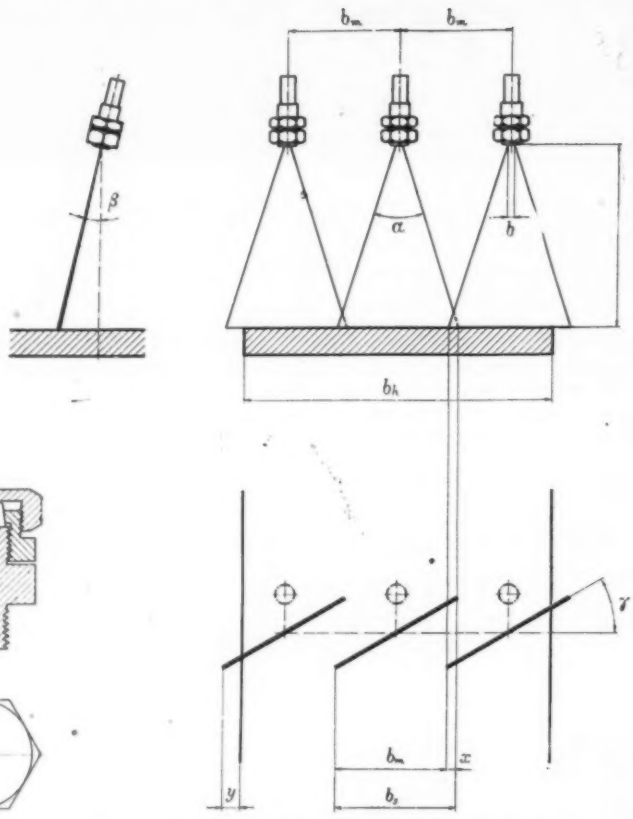

Fig. 3.—Position of the nozzles in relation to rolled material.

Thus tests showed that water-jet cleaning was not particularly advantageous in this case.

A number of tests were carried out on strip rolling. These showed that only a slight improvement was produced by spraying. No general conclusions can, however, be given in view of the effect of different types of rolling, and in some cases a definite improvement was observed with the deep localised rolled-in oxide inclusions which often cause a high percentage of rejections. In the case of stainless steel with 0·05% C, 18% Cr, and 8% Ni the water-spray method is useful, since in rolling to wide strip it is not possible to give heavy reductions.

Effect of Sulphur and Antimony on Steam or Valve Bronze Castings

By A. J. Smith and J. W. Bolton

LOSS of sources for certain high quality, low-impurity virgin metals and diminishing sources of supply for higher purity scrap make it advisable to give very careful consideration to the possible effects of increased "impurities." For alloys used at elevated temperatures the effect of impurities at such temperatures as well as their influence on properties at normal temperatures, as determined in acceptance tests, should be ascertained. These effects should also be determined in the case of substitute alloys.

An investigation was carried out on certain effects of sulphur and antimony, individually and combined on steam or valve bronze containing a minimum of 86·5% of copper, 5·5 to 6·5% of tin, 2 to 5% of zinc, 1 to 2% of lead, 0·05% of phosphorus, 1% of nickel, and 0·25% of iron. (A.S.T.M. Specification, B. 61–41T; Federal Specification, Q.Q.–B–691; Navy Specification, M–46 B.8g.). The first two ingot specifications limit sulphur to 0·05% maximum and antimony to 0·20% maximum, and the Navy specification prescribes manufacture from the highest grade virgin metals and approved scrap, thus implying that the limits of impurities should be quite low.

Alloys were prepared from 40% virgin metals, 60% impurity-free scrap, and were melted in a silica crucible in an induction furnace. Melting and pouring temperatures were 1,315° and 1,175° C. respectively, and the alloys

Bulletin American Society Testing Materials, 1942, No. 116, pp. 18-22.

were cast in green sand moulds used to produce a standard test specimen. Tensile tests at normal temperature and at 288° C. (550° F.) were carried out according to A.S.T.M. standard methods, yield-points being determined by means of an extensometer.

Sulphur was added up to 0·3% or approximately six times that allowed in the ingot specification in powdered form at the beginning of the melt. Recovery was nearly 100%. At normal temperature, average values of 8·35 and 18·3 tons per sq. in. and 41% were obtained for the yield-point, ultimate tensile strength and elongation respectively. At 288° C. the average value obtained for the same properties were 6·05 and 12·3 tons per sq. in. and 18%. The results obtained also showed that at normal temperature and at 288° C. sulphur up to 0·3% produced no adverse effect. Somewhat similar values were obtained for the yield-point, ultimate tensile strength and elongation both at normal temperature and at 288° C. when antimony was added in amounts up to 0·5%. The results obtained showed that this amount of antimony, twice that allowed in the ingot specification, was not detrimental to the physical properties at either normal or elevated temperature.

TABLE I.
COMBINED EFFECT OF SULPHUR AND ANTIMONY ON PHYSICAL PROPERTIES.

Sulphur, %.	Antimony, %.	Tensile Strength Tons per Sq. In.	Elongation, %.
Normal Temperature.			
Nil	Nil	18·25	50·8
0·10	0·15	18·50	52·4
0·05	0·15	18·65	48·5
0·05	0·20	18·45	47·9
0·10	0·15	18·75	47·2
0·10	0·20	18·90	45·8
0·15	0·45	18·85	40·9
0·10	0·55	18·80	40·2
0·15	0·55	18·55	40·0
0·20	0·30	18·45	37·9
0·15	0·20	18·40	36·8
0·15	0·15	17·40	36·3
0·10	0·45	17·90	35·7
0·15	0·50	18·00	34·7
0·20	0·35	17·70	33·8
0·20	0·30	17·85	33·4
0·20	0·55	17·35	30·9
0·25	0·35	17·10	26·8
288° C. (550° F.)			
Nil	Nil	13·65	22·4
0·05	0·15	13·45	24·9
0·05	0·10	12·55	19·8
0·10	0·20	12·75	18·7
0·15	0·15	12·70	17·8
0·20	0·25	12·55	17·5
0·15	0·20	12·20	16·9
0·20	0·30	12·30	16·1
0·15	0·10	12·20	15·6
0·07	0·15	10·90	15·6
0·15	0·30	11·45	14·3
0·20	0·20	11·90	13·9
0·15	0·40	11·45	12·9
0·20	0·50	11·35	12·9
0·10	0·50	11·70	12·8
0·20	0·40	11·55	11·8
0·15	0·50	11·35	11·0
0·25	0·50	11·10	10·9

The combined effect of sulphur and antimony was investigated, sulphur being added up to 0·25% and antimony to 0·55%. The results obtained at normal temperature and at 288° C. are given in Table I, arranged in order of decreasing elongation, together with the tensile strength and nominal sulphur and antimony contents. Examination of the data shows that with increase in sulphur and antimony there is a slight but general decrease in the physical properties and that the apparent limits on sulphur and antimony are 0·15% sulphur and 0·30% antimony. Inferior results were obtained with 0·2% sulphur if the antimony was increased beyond 0·3%. A more marked falling off in physical properties with increase sulphur and antimony was found at 288° C. than at normal temperature, particularly as regards elongation. It was, therefore, considered that if the maximum safe limit for impurities of 18 tons per sq. in. ultimate tensile strength and 40% elongation was applied at normal temperatures, a tensile strength of 9 tons per sq. in., and an elongation of 15% should be attained at 288° C. with good foundry practice.

Tests under foundry operating conditions were not made, but the strength of the metals and their fluidity indicated that density troubles were not likely to occur, and it was felt that changes in characteristics produced by sulphur and antimony additions would have a minor effect in comparison with the major influences of melting, pouring and gating practices.

In general, it was shown that sulphur and antimony contents somewhat higher than those permitted or implied in casting specifications are not detrimental to physical properties at normal temperature or at the maximum permissible service—that of 550° F. (288° C.). It is considered that the maximum impurity limits allowed on these two elements might be revised upwards to 0·15% sulphur and 0·25% antimony, and this would permit greater latitude in the selection of materials, both virgin metals and scrap, during the war.

The Study of Slime Sedimentation

By Sture Mörtsell and Torsten Jeusfelt

THE phenomena associated with slime sedimentation in the so-called "siliceous ore" waste have been studied as a continuation of an earlier investigation by one of the authors.[1]

In order to estimate the course of sedimentation, both the rate of precipitation and the slime content of the clarified liquid were determined. The rate of sedimentation was, to a large extent, recalculated to the so-called "sedimentation index"—i.e., the time in seconds for the main slime boundary to fall 1 cm.

In the different slime sedimentation experiments, measurements were also made of temperature, pH, and conductivity, the last-named expressing the amount of electrolyte in the slime liquid. The samples investigated were taken at different times and treated and stored under different conditions.

By the study of the sedimentation curves for the different experiments—i.e., the position of the main slime boundary plotted as a function of the time, it has been established that the rate of sinking for slime concentrations lower than the critical concentration remains constant within rather narrow limits, which agrees fully with the results of earlier investigations. With regard to the course of the sedimentation curve below the critical point it has previously been stated that at concentrations greater than the critical concentration the rate of sinking is approximately proportional to the height of the slime column. A closer study of this question has shown that the rate of sedimentation in this case is almost certainly proportional, not to the total height of the slime column, but to the height of the slime column above the point to which the slime is packed at the completion of sedimentation.

If the ultimate height of the slime is indicated by z_1, the height of the slime column itself by z, and the time

From *Jernkontorets Annaler*, 1943, **127**, Nos. 6 and 7.

1 S. Mörtsell. "Anordningar för rening av arfalls" vatten Cran anrikningswerk," *Jernkontorets Annaler*, 1940, **124**, No. 8.

by t, the sedimentation curve can be expressed in the following way:—

$$ln\,(z - z_1) = C_1 . t + C_2$$

where C_1 and C_2 are constants (Fig. 1).

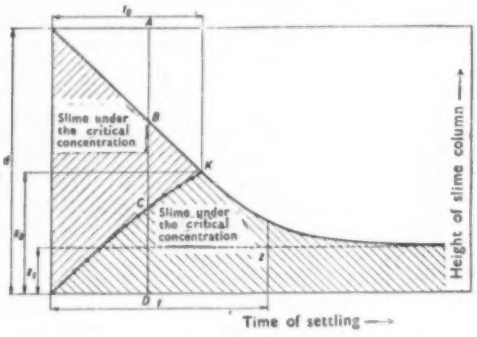
Slime under the critical concentration
Slime under the critical concentration
Height of slime column
Time of settling

Fig. 1.—Slime sedimentation curve: general case.

The layer of liquid above the main slime boundary is referred to as clarified liquid. For those cases in which the sedimentation of the slime took place without any marked main slime boundary—i.e., without much flocculation, the uppermost layer of liquid is referred to as "clarified liquid" (within quotation marks). The clarified liquid is said to undergo *subsequent clarification* with time, whereas the corresponding change in the "clarified liquid" is referred to as *thinning*. From the experiments it has been found that the slime content of the clarified liquid is approximately inversely proportional to the period of subsequent clarification. The course of the *thinning* curve, which is naturally connected with the slime *analysis* of the sample, has not been studied in detail.

The experiments which have been carried out for the investigation of the influence of varying diameter and height of the slime column have shown that in the main the small variations of this kind which occur do not diminish the comparability of the different experiments in other respects. At greater differences in height results have been obtained which suggest that the conditions of clarification may possibly be more favourable in a low slime column than in a high one.

The dependence of the course of sedimentation upon the initial slime concentration has been studied in detail, partly because variations in slime concentration were found to have great practical importance for sedimentation problems as a whole, and partly because in comparable experiments it was found necessary to make corrections even for small changes in slime concentration.

With regard to the *sedimentation index*, experiments have shown that

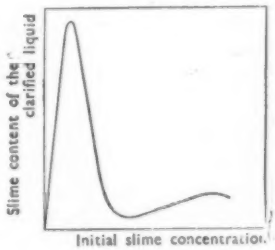

Fig. 2.—General curve, showing slime content of the clarified liquid as function of initial slime concentration.

it undergoes a relatively slow increase at low concentrations, while the increase becomes more rapid at high concentrations. If the initial slime concentration is Q and the sedimentation index I, then the connection between Q and I can be expressed in the general form—

$$ln\,I = C_1\,Q + C_2$$

where C_1 and C_2 are constants, different for different slimes and also for different sedimentation conditions.

The variations of the *slime content of the clarified liquid* in relation to the initial slime concentration are of especial interest. The experimental results have not only verified that two different processes can take place, as has been shown previously (loc. cit.), but have also revealed that the *graph connecting the slime content of the clarified liquid with the initial slime concentration usually shows a minimum*, implying that the two different processes can occur with one and the same slime within different concentration intervals. At low concentrations the clarified liquid is purer at greater concentrations, whereas at high concentrations the slime content in the clarified liquid usually increases with increasing concentrations. For the investigated slimes the minimum, and thus the purest clarified liquid, was reached at initial concentrations of about 50–100 g./litre. The general course of the curve for the clarified liquid is shown in Fig. 2.

With regard to the variation of the *thinning* with the initial slime concentration, the experiments have shown that there is a proportional relationship between the slime content of the "clarified liquid" and the initial concentration.

For "siliceous ore" slime with small contents of electrolytes transitions have been found between clarified liquid and thinning curves for varying slime concentrations.

The effect of differences of grain size in a slime has also been studied, partly with respect to the rate of sedimentation and partly with respect to the slime content of the clarified liquid. For a slime with concentration 100 g./litre and a sedimentation index of, say, 50 sec./cm. the sedimentation index increases by about 0·3 sec./cm. for each increase of 1% in material < 0·060 mm. The larger grain sizes appear to have little or no influence on the nature of the clarified liquid, the concentration of material < 0·060 mm. being evidently the decisive factor.

The study of the influence of temperature on sedimentation has shown that the rate of sedimentation increases rectilinearly with the temperature, in accordance with earlier reports.[2] On this account a correction formula covering differences in temperature is given for the calculation of the sedimentation index. With genuine sedimentation the influence of temperature on the nature of the clarified liquid appears however to be insignificant, whereas the "clarified liquid" becomes purer at higher temperatures because the separate particles sink more rapidly as the viscosity of the water decreases.

The influence of the pH has also been studied, but no definite rules could be established. *pH values in the region 4·5–5·5 and around 9* appear, however, to be favourable both for sedimentation index and for the nature of the clarified liquid.

With regard to the influence of different electrolytes on the course of sedimentation, it has been found that *the slime itself can dissolve in water* to a relatively considerable extent, and that the electrolytes formed in this way can be sufficiently concentrated to have a *noticeable influence on the sedimentation*. This feature explains the inherently favourable sedimentation of certain slimes. The slime itself can give an acid, neutral, or basic reaction in distilled water. For a basic sludge a basic precipitant—e.g., lime—should generally be most suitable, whereas an acid precipitant—e.g.,

2 O. C. Ralston. "The Control of Ore Slimes" *Eng. and Min. J.*, 1916, **101**, 763.

sulphuric acid—should presumably give best results with an acid slime.

The sedimentation properties of a given slime sample usually alter with time. This property has been called *ageing*. Ageing depends partly on the solubility of the slime in water, and partly on certain alterations which can take place slowly even in a dry slime.

Nickel in Bronze

By A. H. Hesse and J. L. Basil

IT has been shown in previous investigations that intercrystalline porosity was considerably reduced when 6% of nickel replaced 6% of tin in a bronze containing approximately 88% of copper, 9% tin, and 3% zinc. This was considered to be extremely important in connection with the manufacture of bronze castings which are required to withstand hydrostatic pressure, because it was felt that the number of rejections due to leakage would be reduced. Although nickel bronzes are used commercially, not too much is known about their internal soundness, or about their mechanical properties and sea-water corrosion resistance, all of which must be considered in the choice of material for many marine applications. A systematic investigation was therefore undertaken to determine the effect of replacing various amounts of tin by nickel on these properties on bronzes of the above composition, on bronze containing approximately 88% copper, 6·5% tin, 4% zinc, and 1·5% lead, and on red brass 85% copper, 5% tin, 5% zinc, and 5% lead.

In the 88–9–3 bronze tin was replaced by 3 and 6% of nickel, in the 88–6·5–4–1·5 bronze by 1 and 2% of nickel, and in the red brass by 2·5% of nickel. The alloys were made from copper, selected scrap, and a copper-nickel alloy, tin and zinc being added where necessary. All charges for castings to be used for density, tensile strength and elongation determinations were melted in a lift coil type of high-frequency induction furnace, while charges for corrosion specimens and fluidity spirals were melted either in the induction furnace or in an oil-fired stationary crucible furnace. Each alloy was cast into three moulds at 1,250° C., 1,150° C., and 1,050° C. respectively.

Since the density of a porous casting is less than the density of a non-porous casting, the degree of porosity may be estimated by a comparison of the densities of the alloy cast under different conditions or with its maximum density. As-cast densities were therefore determined by the conventional displacement of water method, and maximum densities were determined by X-ray diffraction with a back reflection camera. To ensure internal soundness and homogeneity with the latter method, 1 in. cubes of the as-cast alloys were reduced by 75% by several cycles of cold compressing and annealing, and these specimens were then machined to 1 in. + 1 in. + 0·25 in., polished and etched before exposure. A comparison of the maximum densities determined by X-ray methods with the densities of the cast alloys indicated a porosity of about 1% for the 88–9–3 bronze, and for bronzes with a similar composition with the tin replaced by 3 and 6% of nickel respectively, over a wide pouring temperature range (1,050° to 1,250° C..) The 88–6·5–4–1·5 bronze and similar bronzes containing 1 and 2% of nickel showed a decrease in density with increase in pouring temperatures. The addition of nickel to the 85–5–5–5 bronze produced a marked decrease in density when the pouring temperature exceeded 1,150° C.

It having been shown that internal soundness might be improved by replacing tin with nickel, it was then necessary to determine, if the modified composition met the minimum mechanical requirement for the two bronzes and the red brass, and mean tensile-strength-and-elongation-pouring temperature curves showed that these requirements could be met. In the two bronzes the progressive replacement of tin by nickel resulted in less variation in properties due to casting temperature, but the mechanical properties were lowered slightly. The mechanical properties of red brass, however, were not lowered on replacing tin by nickel.

Fluidity curves for the various alloys were determined by plotting spiral lengths in inches against five different casting temperatures between 1,100° and 1,300° C. Those curves showed that the replacement of 6% of tin by 6% of nickel in 88–9–3 bronze markedly lowered the fluidity, while the replacement of 2·5% of tin by 2·5% of nickel in the 88–5–5–5 alloy also lowered the fluidity. Other additions or replacements of tin by nickel in bronze had no marked influence on fluidity. For the two alloys with lower fluidities a higher casting temperature than is normal for the basic compositions is required.

As all the bronzes investigated are well known for their excellent resistance to sea-water corrosion, it was considered necessary, when a change in composition to improve internal soundness was considered, to determine what effect such changes would have on sea-water corrosion as well as on mechanical properties. A series of alloys were therefore tested for resistance to corrosion in the brackish water of the Severn River, at the U.S. Naval Engineering Station, Annapolis, Maryland. The test specimens were rectangular plates and hollow cylinders with finely machined surfaces, and were completely submerged in water. The rectangular specimens were tested at zero velocity and at 30 ft. per second, and the cylindrical specimens at 15 ft. per sec. The modified 88–9–3 bronze containing 6% of nickel was tested in the as-cast, aged and annealed conditions. Aged specimens were water quenched after 6 hours at 760° C. and aged 6 days at 370° C., and the annealed specimens were furnace-cooled after 6 hours at 760° C. All other alloys were tested in the as-cast condition. Tests were carried out on motionless rectangular specimens, rotating rectangular specimens and rotating cylindrical specimens.

The results of these tests showed that the amount of tin that could be replaced by nickel depended on corrosion conditions. The corrosion rate of the motionless rectangular specimens of the 88–9–3 bronze in Severn River water was not affected by chemical composition, while the corrosion rate of the rectangular specimens rotated at 30 ft. per sec., increased as the tin content decreased, and was replaced by nickel and a replacement of 3% tin by nickel appeared to be a maximum for this abnormally high velocity. The corrosion rate of the cylindrical specimens, rotated at 15 ft. per sec., increased but little as the tin content decreased and was replaced by nickel, and no marked composition limit was indicated in this test. The corrosion rates of the alloy containing 6% nickel, 3% tin were not appreciably affected by heat-treatment in the first two tests.

The corrosion rate of the motionless rectangular specimens in River Severn water was not affected by variations

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in the chemical composition of the 88–6·5–4–1·5 bronze, but showed a definite, although slight, increase when 2·5% tin was replaced by nickel in the 85–5–5–5 alloy. The corrosion rate of the rectangular specimens rotated at 30 ft. per sec., definitely increased when 3·5% of nickel replaced the same amount of tin in the former bronze and when 2·5% of tin was replaced by nickel in the latter bronze. A noticeable but slight increase in the corrosion resistance of the cylindrical specimens was noted for the last two mentioned alloys when they were rotated at 15 ft. per sec.

The character of the corrosion product which formed on the motionless rectangular specimens was closely related to the composition of the alloy. Specimens containing no nickel were covered with a thin, tenacious and protective film which changed to a thin scale of low tenacity as nickel replaced tin. Similarly, in the rotating cylindrical tests, the alloys with the lower corrosion resistance were covered by a smooth, glazed film, while those with the higher resistance were covered with a coarse dull film. The least tenacious film formed on specimens of the 88–9–3 bronze containing 6% of nickel.

In general, it was found that the replacement of tin by nickel decreased intercrystalline shrinkage porosity in 88–9–3 bronze, but did not appreciably decrease porosity in the 88–6·5–4–1·5 and 88–5–5–5 bronzes. The replacement of tin by nickel did not appreciably lower the physical properties of the basic compositions, and the extent to which tin was replaced by nickel in all three alloys for marine applications was determined by the corrosion conditions.

The Physical Properties of Steel Wire as Affected by Variations in the Drawing Operations

By H. J. Godfrey

THE purpose of this investigation was to study in detail the physical properties of wire drawn according to a number of drafting schedules and to correlate the physical properties as determined by various types of test methods, so that the method of wire-drawing, resulting in the best combination of physical properties, could then be established. Two materials were used, a 0·67% and a 0·27% carbon steel. All of the 0·67% carbon steel wire came from one coil of ¼ in. diameter hot-rolled rod which was drawn to 0·218 in. in diameter and double lead patented at 840° to 480° C. This patented wire was then drawn 40, 30, 20 and 10% reduction per draft. The 0·27% carbon steel wire was drawn 40, 30 and 20% reduction per draft from a 0·220 in. diameter patented steel rod. A number of sizes of wire representing each rate of drawing were furnished by the wire mill in both coils and machine straightened. Both steels after heat-treatment showed a distinct decarburised layer at their surfaces.

The physical tests carried out included tensile, torsion, bending and bending fatigue. The torsion tests[1] were made, using a gauge length equal to 100 wire diameters and a speed of testing of 60 twists per min., with a tension on the specimen equal to approximately 1% of the tensile strength. Bending tests on the 0·67% and 0·27% carbon steel wires were carried out on two special types of bending machines respectively, and fatigue tests were made on two wire fatigue-testing machines designed on the principle of a rotating wire strut. As the tensile properties of the wire used for the fatigue tests were changed by the machine-straightening operation, and as specimens for the fatigue tests had to be straightened before testing, a number of tests were made, comparing the fatigue properties of machine and hand-straightened wire, and in all cases the results obtained were within the normal variation of the material. All fatigue tests were therefore determined on machine-straightened wire.

The method used in straightening wire was found to have a marked effect on the tensile properties. The tensile strength of the machine-straightened wire was less than that of the hand-straightened material, but the ductility as measured by the elongation was improved. This effect was more pronounced on larger diameter wire, and the loss in strength due to the machine-straightening process diminished as the size of the wire decreased. The effect of the straightening process appeared to indicate that the loss in strength and increase in ductility might be due to the reverse bending which takes place in the rotary machine straightener, and since the endurance limit was only a relatively small percentage of the tensile strength it was not affected by the loss of strength in the machine-straightened wire. In Table I are given the tensile strengths and endurance limits of hand and machine-straightened wire for different diameters.

TABLE I.

TENSILE STRENGTH AND ENDURANCE LIMIT OF HAND AND MACHINE STRAIGHTENED WIRE.

Material.	Diameter, In.	Tensile Strength, Tons per Sq. In.		Endurance Limit, Tons per Sq. In.	
		Hand.	Machine.	Hand.	Machine.
0·67% C Steel.	0·218	71·9	70·5	15·9	16·5
	0·169	90·0	82·6	23·2	21·9
	0·1315	102·7	96·4	26·3	26·8
0·27% C Steel.	0·1025	63·8	59·4	23·2	24·1
	0·079	73·2	68·3	25·0	25·9
	0·0615	81·0	78·1	27·7	27·7
	0·0468	94·2	92·0	29·0	20·4
	0·0368	109·0	105·7	32·1	33·0

The tensile strength of steel wires was in general higher as the rate of drawing increased and the rate of increase in tensile strength was greatest in the latter stages of wire-drawing. This is probably due to the higher rates of drawing having a greater effect on the total cross-section of the wire and resulting in a greater strain-hardening. The amount of heat generated in the die would also be greater for the higher rates of drawing, and this increase in temperature would tend to raise the tensile strength of the wire. Temperature would also account for the small effect of the rate of drawing on the tensile strength of the low carbon steel wire as compared to the higher carbon steel wire, it having been found that up to approximately 250° C. a 0·66% carbon cold-drawn steel wire will increase in tensile strength.

The most accurate method of representing the amount of wire-drawing is probably by true strain, and using this as a measure, the tensile strength after the first reduction is found to increase in direct proportion to the amount of wire-drawing. If this straight-line relationship holds true for all cold-drawn wire, it would be quite advantageous to use the relationship to

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determine the amount and the method of drawing to obtain the necessary tensile strength of a finished wire. The change in the rate of increase in the tensile strength in the final states of wire-drawing is explained by the increase in the amount of preferred orientation, and since the crystal orientation is such that the face diagonal of the cube lies parallel to the wire axis, a change in the rate of increase in strength is to be expected.

The reduction of area as determined by the tensile test increased as the rate of drawing decreased on the 0·67% carbon steel wire, but with the 0·27% carbon steel wire this property was not affected by the rate of drawing. The elongation on 10 in. was slightly improved by a low rate of drawing. It is considered that these changes in ductility might be explained by the temperature effect during wire-drawing. The increase in ductility with a decrease in the rate of drawing might also be due to the more uniform deformation which takes place within the wire-drawing die.

The results obtained on the torsion tests showed that except for one specimen all specimens of the 0·67% carbon steel wire had at least 30 torsions in 100 wire diameters, and the 0·27% carbon steel wire 20 torsions in 100 wire diameters. The rate of drawing and the amount of wire-drawing did not have any consistent effect on the torsion test results. It is probable that this test is affected by the general fibreing of the grains and the surface condition of the wire. The decarburised surface of the material may also have a decided effect on the torsional properties.

The bending properties were considerably improved by reducing the rate of drawing, and were found to be more sensitive to the rate of drawing than any other measure of the physical characteristics of steel wire, and it is considered that this improvement may be due to a more uniform fibreing of the grains as a whole. Since the number of bends are found to decrease after approximately 72% reduction, it is indicated that the beneficial effect of the initial fibreing is being offset by the loss in ductility and by the directional properties which are characteristic of materials which have excess preferred orientation. The bend test also indicates that there is an optimum amount of wire-drawing which will result in a maximum toughness for cold drawn wire, and that the number of bends to cause fracture and the appearance of the fracture obtained are believed to be a measure of the toughness of steel wire.

The rate of drawing effects the endurance limit in the same manner as the tensile strength, and the results show that for the same tensile strength the lower carbon steel has a considerably higher endurance limit. For an endurance limit of 27 tons per sq. in. the tensile strength of the 0·27% carbon steel was only 71 tons per sq. in. as compared to 100 tons per sq. in. for the 0·67% carbon steel wire, even although both materials were definitely carburised. The fatigue properties of both steel wires, however, were considerably lower than would be found on wire free from decarburisation due to the fact that the maximum bending stresses on the fatigue specimen occur at the surface of the wire where the strength of the wire due to decarburisation is at a minimum. The results obtained also showed the endurance ratio to increase slightly up to approximately 80% reduction by wire-drawing and then to decrease rapidly with further wire-drawing.

In general, the best combination of strength, ductility, toughness and fatigue properties resulted from a total amount of wire-drawing of between 75 and 80% reduction.

A New Soft Solder

FOUR research workers in the Division of Metallography at the University of Minnesota have recently published the results of their work carried out to discover a low-cost tinless solder. Considerable publicity has been given to a 97·5% lead–2·5% silver alloy, but the cost of this solder is too high and it has far too high a melting-point for most uses.

The alloy developed by the present work is one of 87·5% lead, 0·5% arsenic, and 12% antimony, which besides being relatively cheap, has better solderability on steel and tinplate than any previous tin-free solders.

Soldering with pure lead is difficult, because of the high melting point and rapid oxidation, whilst the addition of arsenic acts as a flux and assists in preventing oxidation, the optimum additions being between 0·5 to 1% arsenic. Alloys of lead with arsenic also have high melting points, however, and this was overcome by the addition of antimony with a marked increase in solderability. After a number of alloys with different proportions of the three metals had been tried, it was found that that mentioned above had the highest lap-joint strength. It was also the most fluid, had the lowest melting point and wet black steel the best. As its composition is very near the antimony-lead eutectic, it possesses a short solidification range and freezes quickly behind the iron, therefore.

This alloy will not wet a copper soldering iron because of the formation of a copper-antimony compound, and it is also not recommended for copper, brass or galvanised iron work, since with these it has almost no adherence or flowability.

Because this solder has a higher melting point than 50–50 tin-lead solder, either a hotter tool or a hotter flame must be used. A soldering copper tool can be "tinned" with this alloy with the use of zinc chloride, though better success has been obtained by pre-coating the tool with a small amount of 30 tin-70 lead solder.

The materials in this new solder cost $8·50 per 100 lb., and it is calculated that it could be made for 10 cents a pound. Its melting point is 478° F. (30–70 solder melts at 496° F.), and the joint strength of six lapped joints averaged 8,300 lb. per sq. in. (50–50 solder, 8,100 lb.).

Metal Progress, **43** (1), pp. 56–59.

Carburising Characteristics of 0·2% Carbon Alloy and Plain Steels

EIGHT common carburising steels were tested in five different commercial solid carburisers. Chromium steels produced cases with extremely high surface carbon content when carburised in four of the carburisers, and one produced low surface penetration. Nickel opposes chromium in its effect on carbon concentration. Contrary to common belief, lower carburising temperatures do not produce cases of higher surface carbon content. A low carbon skin, about 0·0003 in. to 0·001 in. deep, was produced after carburising for 10 hours or longer at 1,700° F. (925° C.), but was not found at lower temperatures.

G. K. Manning. *Trans. Am. Soc. Metals*, **31** (1), 8 (1943).

Cast Dies for Forging Shells

By P. Attenborough

UNDER present conditions the size of shell forgings range from 1½ in. diameter, weighing 2½ lb., to those of 16 in. diameter, weighing over 2,000 lb. New methods of warfare have increased tremendously the demand for smaller calibre shells, of 5 in. diameter and under, for use by mobile armoured divisions, anti-aircraft guns and small guns mounted on long-range bombing planes. For these sizes of shells considerable success has been attained with the use of cast dies of Mechanite iron for forging the projectiles. One of the most interesting applications of Mechanite dies is in the severe operation of horizontal nose forging, and although this development is recent and somewhat experimental, very promising results have been attained.

Prior to forging operations, shell billets are heated to the required forging temperature, 1,175° to 1,260° C. in either batch or box-type gas or oil-fired furnaces; porthole gas or oil-fired furnaces; continuous furnaces, either the straight-line type or the circular rotary hearth type; or electric furnaces, either high-frequency induction or electrically heated salt bath. The preheating operation is most important in its effect on die life. Improper heating means a greater load on the die and reduces its service life, and uneven heating tends to force the piercing punch towards one side of the shell, throwing the shell forging out of concentricity and gives more intense wear locally on one side of the die. Abnormal loss of heat in transfer of billet from furnace to die also favours excessive die wear, and excessive production of scale due to an oxidising furnace atmosphere may prove detrimental to dies and tools. Following preheating, descaling of billets is effected either mechanically by rollers or cogs, or hydraulically by directing several streams of high-pressure water on to the surface of the billet. The efficiency of the operation has a great influence on die life.

In forging shells three basic methods are used, piercing and drawing, upsetting, and piercing and rolling. In piercing, cast Mechanite dies have been used for some considerable time and have given good results. Actual service results have shown that in piercing 25-pounder shell, cast Mechanite dies of Type G.M. have an average life of 18,000 shells, as against 11,000 for air-hardened nickel-chromium-molybdenum steel. This most widely used type of die made in Mechanite, however, is that used for shell-nosing, the forging operation whereby the final nose shape is given to the clean, machined shell forging.

A summary of the factors affecting die life for both piercing and nose-forming are: (1) Grade of material—Type G.M. (the alloyed Mechanite) was most successful. (2) Correct foundry practice in the placing of gates and runners to give absolute solidity, in the comparative heavy sections obtained, with the type of Mechanite used is essential. (3) Premature die failure can arise from improper fitting of the die to bolster. It is therefore important to fit the die squarely and tightly enough so that the load is taken by the bolster, otherwise dies may fail by bursting. The die should also be concentric with the plunger or shell-forming tool. (4) As the shell or bomb size increases the residual heat in the die increases and the service conditions become more arduous. With existing shell forging equipment no piercing dies for shells have been made for calibres larger than 4·5 in., and nosing dies have usually been limited to shells of 3·45 to 4·5 in. diameter. Bomb dies made in Mechanite may be slightly larger. (5) The greater the degree of deformation in piercing, die life will be decreased in proportion. (6) The control of the preheating treatment of billet before piercing, or pre-heating shells being forged is important as is also the speed of handling after heating. (7) Treatment of dies in use either by cooling with a water spray, by frequent dressing to promote easier flow of the shell form, by frequency of operation or by water cooling. Mechanite die life with water-cooled jackets was 3,000 to 5,000 shells, and with uncooled dies 2,000 to 3,000 shells.

Metals and Alloys, 1943, **18**, No. 2, 287-291.

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F. Catlin in "Fluorescent Method Detects Leaks in Vessels," *Chem. and Met. Eng.*, **50** (8), 1943, p. 116.

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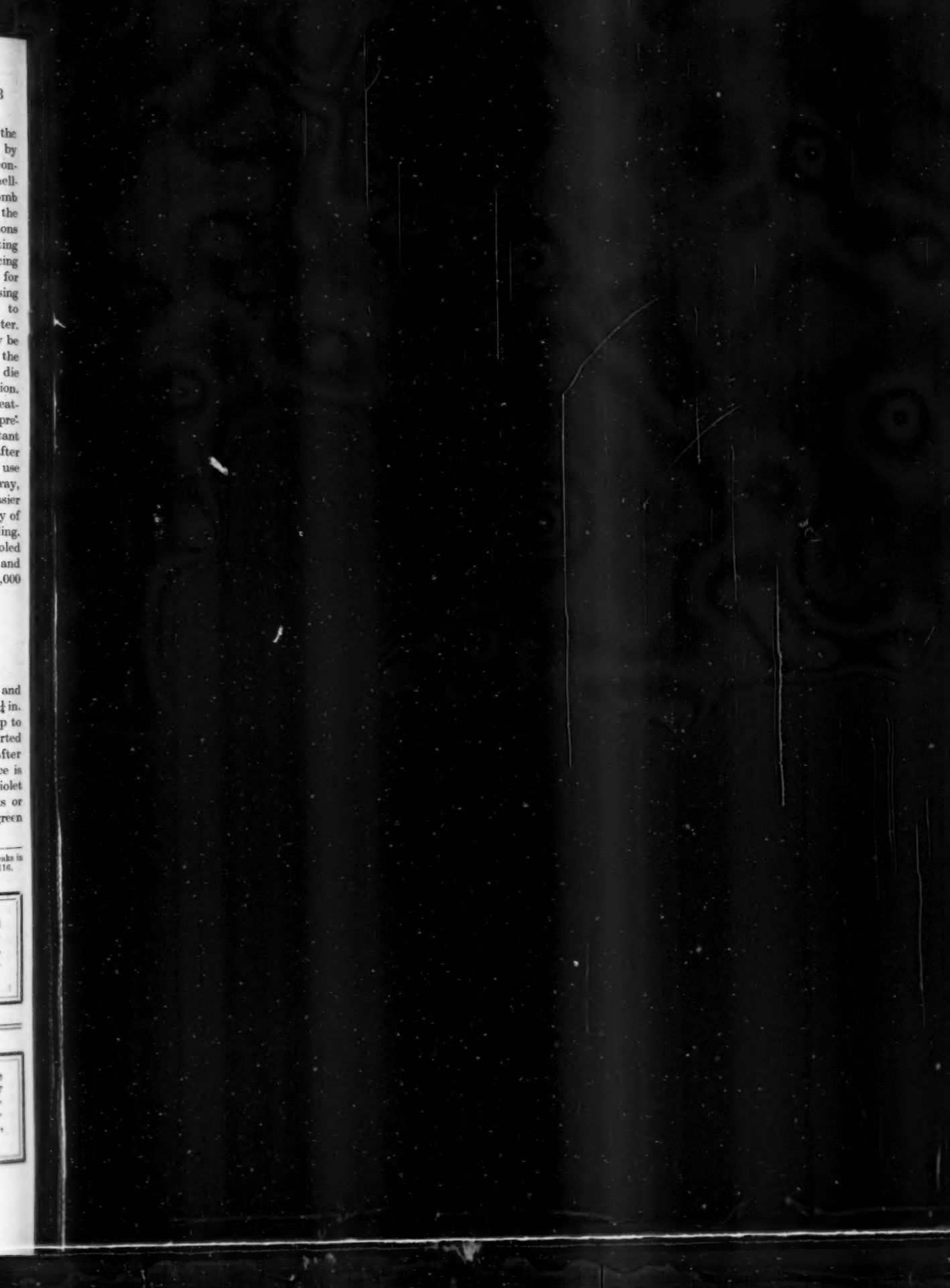
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